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TIME-DEPENDENT EDGE-NOTCH SENSITIVITY OF INCONEL 718 SHEET IN THE TEMPERATURE RANGE 900^{0} TO 1400^{0} F (482^{0} TO 760^{0} C)

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16. Abstract

Time-dependent notch sensitivity of Inconel 718 sheet occurred at 900° to 1200° F (482° to 649° C) when notched specimens were loaded below the yield strength, and tests on smooth specimens showed that small amounts of creep consumed large fractions of creep-rupture life. The severity of the notch sensitivity decreased with decreasing solution treatment temperature and increasing time and/or temperature of the aging treatment. Elimination of the notch sensitivity was correlated with a change in the dislocation mechanism from shearing to by-passing precipitate particles.

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SUMMARY

A study is being made of the severe time-dependent edge-notch sensitivity known to occur at 900° to 1300°F (482° — 704°C) for superalloy sheet materials. An investigation utilizing Waspaloy, to a great extent, established the scope and cause of the problem (ref. 1). Presently reported are results of experiments primarily directed at determining whether the concepts developed also apply to Inconel 718.

Heat treatment variations of 0.030-inch (.75mm) thick Inconel 718 sheet were used to provide a range of mechanical characteristics and microstructural features. Tensile and creep-rupture tests were conducted at temperatures from 900° to 1400°F (482 -760°C). The microstructural features were evaluated for the as-heat treated material and for tested specimens.

Most important, the results showed that the time-dependent notch sensitivity of Inconel 718 could be correlated to the same mechanical characteristics and similar microstructural features as evident for Waspaloy. This would suggest even wider applicability of the results.

Necessary conditions for time-dependent notch sensitivity were (i) the notched specimen loads had to be below the approximate 0.2 percent smooth specimen offset yield strength; and (ii) test data from smooth specimens had to indicate that small amounts of creep used up large fractions of rupture life.

Time-dependent notch sensitivity was observed at test temperatures from 900° to 1200°F (482 -649°C). Decreasing the solution temperature or increasing the time and/or temperature of the aging treatment decreased the susceptibility to time-dependent notch sensitivity.

Variations in heat treatment and test conditions influenced the dislocation motion mechanism. Ni_3Cb(bct) particles (and gamma prime) smaller than a "critical size" were sheared by dislocations. This gave rise to localized deformation and time-dependent notch sensitive behavior. Larger particles were by-passed by the dislocations and the

deformation was homogeneous. Under these conditions, no time-dependent notch sensitivity was observed.

INTRODUCTION

The results presented were derived from a current study being made of the severe time-dependent edge-notch sensitivity that has been observed for nickel-base superalloy sheet materials at temperatures from 900° to 1300°F (482 - 704°C). The research was carried out at The University of Michigan, Ann Arbor, Michigan, under sponsorship of the National Aeronautics and Space Administration.

Extensive research (ref. 1) established the scope and the cause of the problem for Waspaloy. In addition, heat treatments were defined which eliminated the time-dependent notch sensitivity. Continuing research is directed at broadening the applicability of the concepts developed for Waspaloy. To achieve this, the study is being extended to include other alloys. The results reported are those obtained for Inconel 718, the composition of which differs considerably from that of Waspaloy. Of major significance, columbium is present in Inconel 718 but not in Waspaloy. This element has a marked influence on phase relations and hence on the microstructures and mechanical behavior.

As part of an evaluation of various superalloys in sheet form for potential air frame skin applications Inconel 718 in three heat treated conditions was shown to exhibit time-dependent notch sensitivity at 1000° and 1200° F (538° and 649° C) (ref. 2). (Representative results from this investigation are included as Figures 2 through 7). The research presently reported was designed to extend the scope of these results. Heat treatments were selected to provide a wide range of microstructural features. These were also expected to produce considerable variation in mechanical characteristics. Tensile and creep-rupture tests were carried out at temperatures from 900° to 1200°F (482 - 649°C) where severe time-dependent notch sensitivity can occur. Testing was also carried out at 1400°F (760°C) in order to provide a contrasting case where the notch sensitivity has not been observed. The microstructural features, particularly the dislocation motion mechanisms in the tested specimens, were evaluated.

EXPERIMENTAL DETAILS

Materials

The commercially produced Inconel 718 used in the investigation had the following reported composition (weight percent):

Ni	C	Mn	Fe	S	Si	Cr	Al	Ti
53.97	0.05	0.12	16.50	0.007	0.22	18.98	0.52	1.04

Co	Mo	Cb	B
0.05	3.15	5.25	0.002

The material was received as 0.030-inch (.75mm) thick cold reduced sheet. Specimen blanks were cut in the longitudinal direction prior to heat treatment.

Heat Treatment

The heat treatments were designed to produce a wide range of microstructural features. The selection was based primarily on isothermal transformation diagrams reported for the alloy (ref. 3). The heat treatments were as follows:

	Solution Treatment		Aging Treatment
1.	10 hours at 1950°F (1066°C)	+	48 hours at 1350°F (732°C)
2.	1 hour at 1950°F (1066°C)	+	48 hours at 1350°F (732°C)
3.	1 hour at 1950°F (1066°C)	+	2 hours at 1550°F (843°C)
4.	1 hour at 1950°F (1066°C)	+	24 hours at 1550°F (843°C)
5.	10 hours at 1800°F (982°C)	+	48 hours at 1350°F (732°C)
6.	10 hours at 1700°F (927°C)	+	3 hours at 1325°F (718°C)
7.	10 hours at 1700°F (927°C)	+	48 hours at 1350°F (732°C)
8.	1 hour at 1700°F (927°C)	+	3 hours at 1325°F (718°C)
9.	1 hour at 1700°F (927°C)	+	2 hours at 1550°F (843°C)

The blanks were solution treated individually in an argon atmosphere and then air cooled. To prevent warping, the blanks were aged while clamped in a fixture in batches of 10 or 12. It should be noted that although the higher temperature exposures are referred throughout this paper as "solution treatments", the use of this designation does not necessarily signify complete solution of all constituent phases.

Testing Procedures

The testing procedures used have been described in depth elsewhere (ref. 1). After heat treatment blanks were machined into smooth and sharp-edge (Kt>20) notched specimens (fig. 1).

The tensile tests were conducted using a hydraulic tensile machine. Smooth specimens were tested at a strain rate of approximately 0.01 per minute up to about 2 percent deformation. The strain rate was then increased to about 0.05 per minute until failure. Notched specimens were loaded at a rate of 1000 psi per second (6.9 MN/m^2 per second). The creep-rupture tests were conducted in beam loaded machines. The rupture times were recorded automatically. For both tensile and creep-rupture tests, recommended practices (ASTM-E21, E139) were followed in control of test temperatures and distributions. The extensions were measured by an optical extension system which has a sensitivity of five millionths of an inch (0.125 μm).

Structural Examination

Conventional methods were employed for microstructural examination. Samples for optical microscopy and replica electron microscopy were etched electrolytically in "G" etch, an etchant developed by Bigelow et al (ref. 4). Samples approximately 0.5 inches wide by 0.7 inches long (1.3 x 1.8 cm) for transmission electron microscopy of the tested specimens were cut from the gauge lengths, ground on wet silicon carbide papers and electropolished. This was carried out at an applied voltage of 20 volts in conjunction with a chilled mixture of 83 percent methanol, 7.5 percent sulphuric acid, 3 percent nitric acid, 4.5 percent lactic acid and 2 percent hydrofluric acid. The thin films were studied and photomicrographed in a JEM electron microscope operated at 100 KV.

X-ray diffraction analysis of extracted residues was used to identify the phases present in the as-heat treated materials. The residues were obtained by preferentially dissolving the matrix electroytically in a solution of 10 percent phosphoric acid in water. A platinum cathode was used at 3-4 volts potential. The residues were washed with alcohol, dried and formed into thin wires using a Duco Cement binder. X-ray exposures were conducted in a 144.6mm diameter Debye camera using nickel-filtered copper ratiation (40KV, 16mA) for a period of four hours. The line positions were measured and the "d" values calculated. The results were then analyzed by comparison with standard patterns available from the literature and from the files of the Joint Committee on Powder Diffraction Studies.

INFLUENCE OF HEAT TREATMENT ON THE MECHANICAL CHARACTERISTICS

The results of tensile and creep rupture tests are presented in Tables 1 through 10. The rupture data are included as stress-rupture time curves and as parameter curves in Figures 8 through 25. The Larson-Miller parameter with C of 20 was used simply because it was an effective method of presenting the data.

The heat treatment variations resulted in a wide range of mechanical characteristics. In particular, the severity of the time-dependent notch sensitivity varied considerably with changes in both the solution and aging treatments. The results for the material heat treated 10 hours at 1950°F (1066°C) plus 48 hours at 1350°F (732°C), provide an example of severe time-dependent notch sensitivity (table 2). At the shorter times at the lower test temperatures, the notched specimen rupture curves were somewhat below those for smooth specimens (figs. 8, 9). The notched to smooth rupture strength ratios (N/S) were the same order as determined by tensile tests (table 2). At varying time periods, the notched specimen rupture curves exhibited drastic increases in steepness so that the N/S ratio decreased with time to values considerably below those obtained in tensile tests, i.e. the material exhibited time-dependent notch sensitivity. At intermediate temperatures, upward breaks in the rupture curve were evident. This resulted in an increase in the N/S ratio or a decrease in notch sensitivity. The rupture strength ratio increased with increasing test temperature and longer rupture times, until, at the highest temperature a value of about 1.0 was obtained.

In contrast to the above behavior, the results for the material heat treated 10 hours at 1700°F (927°C) plus 48 hours at 1350°F (732°C) are typical of the heat treated conditions that were not susceptible to time-dependent notch sensitivity (table 8, figs. 20, 21). The N/S rupture strength ratios were generally higher than those obtained in tensile tests.

The types of characteristics described above were evident for the other heat treated materials (figs. 10 through 19 and 22 through 25). The severity of the time-dependent notch sensitivity for the heat treated conditions studied, including those previously reported (ref. 2), are tabulated below. Ratings of 2 and 1 correspond to severe and limited time-dependent notch sensitivity respectively. Test conditions for which no time-dependent notch sensitivity was evident were rated 0.

		Test Temperature				
	°F	900	1000	1100	1200	1400
Solution Treatment	Aging Treatments °C	482	538	593	649	760
10 h. 1950°F(1066°C)	+ 48 h. 1350°F(732°C)	2	2	2	2	0
1 h. 1950°F(1066°C)	+ 48 h. 1350°F(732°C)	2	2	2	2	0
	+ 2 h. 1550°F(843°C)	2	2	2	1	0
	+ 24 h. 1550°F(843°C)		0	0	?	0
10 h. 1800°F(982°C)	+ 48 h. 1350°F(732°C)	2	2	1	?	0
10 h. 1700°F(927°C)	+ 3 h. 1325°F(718°C)	2	1	0	0	0
	+ 48 h. 1350°F(732°C)	0	0	0	0	0
1 h. 1700°F(927°C)	+ 3 h. 1325°F(718°C)	2	2	?	0	0
	+ 2 h. 1550°F(843°C)		0	0	0	0
Cold Worked 20%	+ "Multiple" 1*		2		2	
1 h. 1950°F(1066°C)	+ "Multiple" 2		2		2	
1 h. 1750°F(954°C)	+ "Multiple" 1	2	2	1	?	

*"Multiple" 1 - 1325°F(718°C)/8 h., F.C. to 1150°F(621°C) in 10 h., A.C.
"Multiple" 2 - 1350°F(732°C)/8 h., F.C. to 1200°F(649°C) in 12 h., A.C.

The principal features evident were as follows:

(1) Decreasing the solution temperature decreased the susceptibility to time-dependent notch sensitivity.

(2) Increasing the severity of the aging treatment (increasing time and/or temperature) reduced the susceptibility to time-dependent notch sensitivity.

(3) The materials with the commonly used "multiple" aging treatments exhibited time-dependent notch sensitivity.

For all heat treated conditions, the rupture strengths decreased rapidly with increasing time and/or temperature for parameter values above about 37 (figs. 9, 11, 13, 15, 17, 19, 21, 23, and 25). Extensive use is made of Inconel 718 for "high temperature" applications for which the parameter values are relatively low. Under these conditions, the smooth specimen rupture strengths are high, however, severe time-dependent notch sensitivity can occur.

At low parameter values, the notched specimen rupture strengths were below those for smooth specimens. The rupture strengths for notched specimens varied considerably with heat treatment. Particularly important was whether or not the material exhibited time-dependent notch sensitivity. Those which did not had similar notched rupture strengths, while those which were notch sensitive had considerably lower strengths (fig. 26). It should also be noted that the commonly used "multiple" aging treatments, evaluated previously (fig. 2 through 7), resulted in lower notched rupture strengths than obtained for a number of the heat treatments used in the present investigation. This occurred even though the heat treatments used were not specifically designed to maximize the notched rupture strengths.

The heat treatments studied did not result in a very wide range of smooth specimen tensile and creep-rupture strengths (table 1 through 10). Differences in rupture strength from variations in solution and aging treatments can best be seen from comparison of the parameter curves, such are presented as Figures 27, 28 and 29.

Rupture ductility has often been used to indicate the susceptibility of a material to notch sensitivity. Results reported for Waspaloy (ref. 1) indicated that no such relationship was generally applicable. The same conclusion was drawn from analysis of the elongation and reduction of area values at rupture for Inconel 718 (tables 2 through 10). For both alloys, the results indicated that factors which contribute to ductility, rather than the ductility per se, control the time-dependent notch sensitivity. This was evident from analysis of the deformation-time characteristics (reported in a later section for Inconel 718).

FRACTURE CHARACTERISTICS

The fracture characteristics of Inconel 718 were similar to those established for Waspaloy (ref. 1). Consequently, this aspect is not reported in depth. Both smooth and notched rupture tested specimens failed by initiation and relatively slow growth of intergranular cracks (fig. 30) followed by transgranular fracture. The intergranular part of the fracture was perpendicular to the loading axis and was dark in color from oxidation. The remainder of the fracture was not appreciably discolored by oxidation and was a typical shear failure slanted through the thickness. This latter fracture occurred when the increase in stress on the load bearing area, due to growth of the intergranular crack, exceeded that necessary to cause rapid shear. In consequence, the lengths of the intergranular cracks (expressed as a percentage of specimen width in tables 2 through 10) increased with decreasing test stress and thus with increasing test time.

The creep deformation that occurred in the smooth specimens was relatively uniform throughout the gauge section. In contrast, the deformation in the notched specimens was localized. Initially dimples (areas of severe deformation) formed at the base of the notches. Subsequently, intergranular cracks initiated in these regions (fig. 30). The dimples remained at the head of the cracks as they progressed across the specimens.

Intergranular cracks were found in notched but not smooth specimen tests discontinued before rupture. This is in agreement with the crack growth rate studies reported for Waspaloy (ref. 1). These showed that visible cracks formed in notched specimens after about 60 percent of the rupture life. The life fraction was smaller than required for crack initiation in smooth specimens (greater than 95%). The results also showed that the occurrence of time-dependent notch sensitivity is due to more rapid initiation of intergranular cracks in notched than in smooth specimens. This was shown to be related to the relaxation by creep of stress concentrations introduced by the presence of the edge-notches.

STRESS RELAXATION

Relaxation of stress concentrations can occur by "yielding" on loading (time-independent deformation) and by subsequent creep (time-dependent deformation). For notched specimen tests loaded above the approximate 0.2 percent offset yield strengths (established by smooth specimens tests) no time-dependent notch sensitivity was observed (figs 8 through 25). This occurred since "yielding" on loading reduced the stresses across the specimens at the base of the notches to approximately the nominal stresses.

For Waspaloy, time-dependent notch sensitivity occurred in notched specimens loaded below their yield strength, when tests of smooth specimens showed that small amounts of creep consumed large fractions of creep-rupture life (ref. 1). In other words, when the creep deformation necessary to relax stress concentrations caused excessive damage resulting in premature initiation of intergranular cracks. The smooth specimen deformation characteristics of Inconel 718 were examined to determine whether a similar correlation existed. Iso-creep strain curves were constructed for each temperature on plots of life fraction versus test stress (figs. 31 through 39). These curves were derived for 0.1 and 0.2 percent creep deformation (the order of deformation necessary to ensure relaxation of elastic stresses from the approximate yield stress to the nominal stresses.) The inclusion of curves for 0.5, 1, and 2 percent creep strain aided establishment of the nature of the curves for lower strain.

Again, a correlation was evident between the time-dependent notch sensitive behavior and the characteristics of the iso-creep strain curves. For the material solution treated 10 hours at 1950°F (1066°C) plus 48 hours at 1350°F (732°C), the life fractions for small amounts of creep strain at 1000°F (538°C) increased drastically as the test stress decreased (fig. 31). For notched tests loaded to normal stresses below about 110 ksi (758MN/m^2), the relaxation of the stresses (from the approximate yield stress to the nominal) would consume considerable if not all of the creep-rupture life of the material at the base of the notch. Thus, as observed experimentally (figs. 8, 9), time-dependent notch sensitivity behavior would be expected. At 1200°F (649°C) {and presumably 1100°F (593°C)} as the test stress decreased, the life fractions for 0.1 and 0.2 percent strain increased to relatively high levels and then subsequently decreased. These results are consistent with the observed increase followed by a decrease in time-dependent notch sensitivity. At 1400°F (760°C) the life fractions were at relatively low levels and no time-dependent notch sensitivity occurred.

The results for the materials heat treated 10 hours at 1700°F (927°C) plus 48 hours at 1350°F (732°C) are in contrast to those presented above, (fig. 37). The life fractions for small amounts of creep strain remained at low levels for all test conditions. In accordance with this, no time-dependent notch sensitivity was observed (figs. 20, 21).

Analysis of the data for the other heat treated materials showed similar correlations between the nature of the iso-creep strain curves (figs. 32 through 39) and the time-dependent notch sensitive behavior (figs. 10 through 25).

MICROSTRUCTURAL FEATURES CONTRIBUTING TO TIME-DEPENDENT NOTCH SENSITIVITY

Many nickel-base superalloys (including Waspaloy) age harden by precipitation of an LI_2 - ordered fcc phase, $Ni_3(Al, Ti)$, called gamma prime. Inconel 718 differs in that age hardening occurs primarily due to precipitation of Ni_3Cb(ref. 5). This phase, designated γ'', has a DO_{22} - ordered bct structure. The γ'' phase is metastable so that it is replaced on thermal exposure by the β phase. This phase is also Ni_3Cb, but it has a Cu_3Ti - ordered orthorhombic structure.

In the study of Waspaloy (ref. 1), a correlation was established between the dislocation motion mechanism operative and the time-dependent notch sensitivity. Dislocations sheared γ' particles smaller than a critical size. Particles larger than the critical were by-passed by dislocations. The former mechanism promoted the deformation-characteristics that gave rise to the time-dependent notch sensitivity. The microstructural features of Inconel 718 were studied to determine whether a similar correlation occurred.

Initially optical metallography, replica and transmission electron microscopy and X-ray diffraction studies were carried out for the as-heat treated materials. Subsequently, smooth specimens which had been creep-rupture tested were examined by transmission election microscopy. The specimens studied were selected from heat treatments exhibiting a range of time-dependent notch sensitive behavior.

Original Microstructures

Typical optical and electron micrographs are presented in Figures 40, 41 and 42. These together with the results of the X-ray diffraction studies (Table 11), were used to characterize the microstructural features in the as-heat treated materials.

Increasing the time of solution treatment at 1950°F (1066°C) from 1 to 10 hours resulted in a considerable increase in grain size. This reflects the absence of restraint or grain growth associated with large amounts of precipitate particles. Ti(C,N) is the only precipitate expected to be present at 1950°F (1066°C). Presumably, these are the large particles evident in the optical micrographs for the material aged 48 hours at 1350°F (732°C) after solution treatment at 1950°F (1066°C) (figs. 40a, b). X-ray diffraction of extracted residues indicated the presence of Cb, Ti(C, N), γ' and/or* γ'' in these materials (Table XI). These latter phases were not readily resolvable by electron microscope replica techniques (fig. 41 a, b). In thin films, (fig. 42 a), the carbide particles observed were primarily present as "plate-like" grain boundary precipitates while the γ' and/or γ'' were intragranular precipitates about 300 Å in diameter. The presence of γ'' was demonstrated by an electron diffraction technique (ref. 6). The method uses a [100] diffraction pattern (fig. 43). Superlattice reflections of the form 1-1/2-0 are allowed for the bct γ'' phase but not for the fcc γ'. Thus, the occurrence of this reflection demonstrated that the heat treated material contained γ''. The presence of γ' was inferred from subsequently reported metallographic observations for materials at higher temperatures. (In cases such as this, where the γ' and the γ'' could not be distinguished visibly, the precipitate will be referred to as γ'/γ''.)

For the materials solution treated at 1950°F (1066°C) and aged 2 and 24 hours at 1550°F (843°C), the presence of γ' and γ'' gave the optical micrographs a mottled appearance (fig. 40 c, d). The particles were large enough to be resolvable using replica techniques (fig. 41 c, d). The γ' was present as spherical particles

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*The γ' and γ'' phases are difficult to distinguish using X-ray diffraction. The "d" values are similar. Differences do occur in the superlattice reflections but these are not readily resolvable (ref. 5).

with a relatively low volume fraction. The average size of the particles was about 450Å and 1100 Å for the 2 and 24 hour treatments respectively. The majority of the precipitate particles were plates of γ'' (precipitates coherently with the c-axis normal to the plane of the plates and along any of the three <110> fcc directions - ref. 5). The approximate average thickness and length of the plates were respectively 200Å and 1000 Å for the 2 hour treatment and 500 Å and 4000 Å for the 24 hour treatment.

X-ray diffraction indicated the presence of small amounts of β phase for the materials aged at 1550°F (843°C) after solution treatment at 1950°F (1066°C) (Table 11). In micrographs, this phase was evident as needles, predominately alongside grain boundaries (figs. 40 d, 41 d). The areas adjacent to grain boundaries and β precipitate particles were depleted of γ'' (fig. 41 d).

The majority of the precipitate present in the optical micrographs of the material heat treated 10 hours at 1800°F (982°C) plus 48 hours at 1350°F (732°C) was β phase (fig. 40 f). This phase formed during the 1800°F (982°C) solution treatment (fig. 40 e). X-ray diffraction showed that the aged material also contained Cb, Ti(C, N), γ' and/or γ''.

All of the materials solution treated at 1700°F (927°C) and aged contained Cb, Ti(C, N), γ'/γ'' and the β phase (Table 11). Ni_3Cb needles precipitated during the 1700°F (927°C) treatments (fig. 40 g, j). A much larger amount of β phase was present after the 10 hour exposure than the 1 hour treatment. Aging 3 hours at 1325°F (718°C) or 48 hours at 1350°F (732°C) after solution treatment at 1700°F (927°C) resulted in γ'/γ'' too small to be resolved using replica techniques (figs. 41 f, g). In thin films (figs. 42 b), resolution was also difficult because the γ'/γ'' particles were only about 60 Å and 200 Å in diameter for the 1325°F (718°C) and 1350°F (732°C) treatments, respectively. These particle sizes are smaller than those produced by similar aging treatments after solution treatment at 1950°F (1066°C). This also occurred for the 2 hour at 1550°F (843°C) aging treatment (fig. 41 c, h).

Microstructures of Tested Specimens

Examination of tested specimens by transmission electron microscopy was carried out primarily to determine the dislocation structures present. The observations were made for selected heat treatments and test conditions. The dislocation arrangements were expected to be representative of all of the heat treatments used in the study.

(1) Material heat treated 1 hour at 1950°F (1066°C) plus 48 hours at 1350°F (732°C):

For the specimen tested at 1100°F (593°C) {at 120 ksi (827MN/m^2) ruptured in 1.4 hours} the most obvious feature was the {111} planar slip banding (fig. 44). This reflects shearing of the γ'/γ'' coherent precipitates by dislocations (refs. 5, 7). Similar dislocation structures would be expected to occur in other specimens tested at temperatures low enough so that little or no growth or γ'/γ'' occurs.

It was evident from microstructures of the specimen tested at 1400°F (760°C) {at 30 ksi (207MN/m^2) ruptured in 384 hours} that structural changes had occurred during the test exposure. Ni_3Cb needles precipitated (fig. 45, 46) and the γ' particles increased in size to about 750 Å. The γ'' also grew so that it was clearly resolvable as plates approximately 500 Å thick and 4000 Å long. Contrast effects associated with coherency (ref. 8) were observed for both γ' and γ'' precipitate particles (fig. 45 b, c). Because of the presence of large precipitates the deformation was homogeneous (fig. 46). Dislocations were observed entangled with the γ'' particles and in some cases forming loops around the γ'. It must be assumed that in the early part of the test when the particles were small, the dislocations sheared the particles, i.e. the microstructure would have been similar to Figure 44.

The above observations are analogous to those reported for aspaloy (ref. 1). In this case, dislocations sheared γ' particles naller than a critical size. When the particles were larger than e critical, they were by-passed by dislocations. These mechanms resulted in localized and homogeneous deformation respectively.

) <u>Material heat treated 1 hour at 1950°F (1066°C) plus 2 hours at 1550°F (843°C)</u>

The deformation that occurred for the specimen tested at 1100°F 93°C) {at 100 ksi (690 MN/m^2) ruptured in 385 hours} was localized . slip bands (fig. 47a). Presumably, dislocations sheared the γ' ırticles (about 450 Å in diameter) and also the γ'' (200 Å thick and)00 Å long). The previously described results would indicate that rowth of the precipitates during higher temperature tests would ıuse the deformation to become homogeneous.

One additional feature was evident from the study of the specimen :sted at 1100°F (593°C). In a number of micrographs, a fine precipitate ıbout 70 Å in diameter) was detected (fig. 47 b). Presumably, this ; γ'/γ'' that formed subsequent to the 1550°F (843°C) aging treatment nd developed during the test exposure.

3) <u>Material heat treated 1 hour at 1950°F (1066°C) plus 24 hours at 1550°F (843°C)</u>

The γ'/γ'' in the aged material was larger than "critical" size. :ven in a low temperature test specimen{at 1000°F (538°C) and 115 ksi 793 MN/m^2), ruptured in 1857 hours} the dislocations were homogeneously istributed (fig. 48).

A dispersion of γ'/γ'' (about 100 Å in diameter) was observed in he specimen tested at 1100°F (593°C) {at 100 ksi (690 MN/m^2) ruptured n 3528 hours}. This feature was similar to that described for the naterial aged 2 hours at 1550°F (843°C).

4) <u>Material heat treated 1 hour at 1700°F (927°C) plus 3 hours at 1325°F (718°C)</u>

The deformation in the specimen tested at 1000°F (538°C) {at 130 :si (896 MN/m^2) ruptured in 5613 hours } was localized (fig. 49 a). In the majority of cases, the dislocations in pile-ups were dissociated to form stacking fault ribbons. This type of deformation was not expected because, in the presence of γ'', it requires coplanar motion of multiple dislocations. Four whole dislocations must move along the same plane to restore order for all three orientations of γ'' (ref. 7). (It should be noted that the presence of γ'' was confirmed by electron diffraction.)

Growth of γ'/γ'' occurred during exposure of the specimen tested at 1200°F (649°C) {at 65 ksi (448 MN/m^2) ruptured in 937 hours}. As a result, the dislocations were homogeneously distributed (fig. 49b). The microstructures indicated that the dislocations bowed between the γ'/γ'' particles (about 250 Å in diameter), leaving pinched off dislocation loops, i.e. the dislocations by-passed rather than sheared the particles. This was probably due to differences in the volume fraction of precipitate. Although not determined as part of the investigation, less γ'' was probably present for materials aged after solution treatment at 1700°F (927°C) than for those treated at higher temperatures, i.e. 1950°F (1066°C). This could be expected because precipitation of Ni_3Cb needles during the 1700°F (927°C) treatment must reduce the amount of Cb in solid solution available to form γ'' during aging. Thus, the volume fraction of the γ'/γ'' precipitate was reduced, which should lower the "critical" size (ref. 9). Further research is necessary to clarify these effects.

(5) <u>Material heat treated 10 hours at 1700°F (927°C) plus 3 hours at 1325°F (718°C).</u>

Limited study of a specimen tested at 1000°F (538°C) {at 130 ksi (896 MN/m^2) ruptured in 391 hours} did not reveal any inconsistencies from the results described above for the material solution-treated 1 hour at 1700°F (927°C) and aged at 1325°F (718°C).

(6) Material heat treated 10 hours at 1700°F (927°C) plus 48 hours at 1350°F (732°C)

For the specimen tested at 1000°F (538°C) {at 120 ksi (827 MN/m^2) ruptured in 1382 hours} the deformation was homogeneous (fig. 50). This result demonstrated that the γ'/γ'' produced by the aging treatment (about 200 Å in diameter) was larger than the "critical" size.

Correlation of the Time-Dependent Notch Sensitivity with Microstructural Features

The results indicate that a correlation exists between the predominant dislocation mechanism and the time-dependent notch sensitivity. The relationship was the same as evident for Waspaloy (ref. 1). Shearing the precipitate particles by dislocations, resulted in greater susceptibility to time-dependent notch sensitivity than when they were by-passed. For the materials heat treated 1 hour at 1950°F (1066°C) plus 48 hours at 1350°F (732°C), 1 hour at 1950°F (1066°C) plus 2 hours at 1550°F (843°C), 1 hour at 1700°F (927°C) plus 3 hours at 1325°F (718°C), and 10 hours at 1700°F (927°C) plus 3 hours at 1325°F (718°C), time dependent notch sensitivity was observed at the lower test temperatures. During these tests, the γ'/γ'' particles were sheared by dislocations and the deformation was localized. During the higher temperature tests, growth of the γ'' and γ' precipitates occurred. This resulted in a change of dislocation motion to "by-passing" so that a homogeneous distribution of dislocations resulted. This correlates with the elimination of time-dependent notch sensitivity by increasing the test temperature.

For the materials heat treated 1 hour at 1950°F (1066°C) plus 24 hours at 1550°F(843°C) and 10 hours at 1700°F (927°C) plus 48 hours at 1350°F (732°C), the dislocations were homogeneously distributed and no time-dependent notch sensitivity was observed.

There was no evidence to indicate that the other heat treated materials, for which tested specimens were not studied, would not follow the above correlation.

It is of interest to compare the behavior of materials with a given aging treatment, e.g. 48 hours at 1350°F (732°C). The time-dependent notch sensitivity was severe for the material solution treated at 1950°F (1066°C). Decreasing the solution temperature to 1800°F (982°C) decreased the notch sensitivity, until for the 1700°F (927°C) treatment, none was observed. This is consistent with the metallographic observation that the "critical" size decreased with decreasing solution temperature. As suggested previously, this probably occured, because lowering the solution temperature reduced the volume fraction of γ'/γ''. This would also explain why the time dependent notch sensitivity of the material heat treated 1 hour at 1700°F (927°C) plus 3 hours at 1325°F (718°C) was more severe, or occurred at higher test temperatures than for the material heat treated 10 hours at 1700°F (927°C) plus 3 hours at 1325°F (718°C).

Failure of both smooth and notched creep-rupture specimens occurred by relatively slow intergranular crack initiation and growth, followed by transgranular fracture. Consequently, factors which affect the intergranular crack initiation influence the rupture times and hence the time-dependent notch sensitive behavior. Intergranular crack initiation must be dependent on (a) the nature of intragranular deformation and (b) the metallurgical characteristics of the grain boundaries. The influence of variations in the grain boundary characteristics on the time-dependent notch sensitivity was not evident from the results. Nor was a relationship evident from the study of Waspaloy (ref. 1). In both cases, a correlation was evident between the dislocation mechanism and the time-dependent notch sensitive behavior. This would suggest that the influence of the γ'' and/or γ' precipitates on the notch sensitivity overshadows effects from variations in grain boundary characteristics.

Hardness Testing

Increasing the particle size of γ' and/or γ'' increases the critical resolved shear stress (CRSS) for dislocations motion until the particle dimensions exceed a critical size, after which, further increase in size decreases the CRSS. Particles below the critical size are sheared by the dislocations, while larger particles are by-passed. Hardness values were obtained for a range of heat treated materials, including those used in the test program, to determine whether these values, by reflecting the CRSS, could be used to monitor the γ'/γ'' size relative to the critical. The results (fig.51) showed the following:

(1) The hardness values for the material heat treated 1 hour at 1950°F (1066°C) plus 24 hours at 1550°F (843°C) indicated that the γ'/γ'' particles were larger than the critical size (fig. 51a). Correspondingly, no time-dependent notch sensitivity was observed.

(2) From results for aging at 1325°F (718°F) and 1400°F (760°C) hardness values were estimated for aging at 1350°F (732°C). These indicated that aging 3 hours at 1350°F (732°C) after solution treatment 1 or 10 hours at 1700°F (927°C) resulted in γ'/γ'' particles smaller than the critical size (fig. 51 b, e). In accordance with this, these materials exhibited time-dependent notch sensitivity.

(3) For many heat treated materials, the hardness values indicated that the γ'/γ'' sizes were near the critical. In these cases, it would not be possible to guarantee correct prediction of the time-dependent notch sensitive behavior. Certainly, for none of the heat treatments evaluated was there a case for which the hardness test results clearly indicated the incorrect notch sensitive behavior.

(4) The aging time at which the maximum in hardness occurred at each temperature decreased as the solution temperature decreased. This corresponds to the observed decrease in "critical size".

(5) The results also indicate that solution treated material aged 8 hours at 1325°F (718 °C) F. C. to 1150 °F

(621°C) in 10 hours A.C. could be expected to be susceptible to time-dependent notch sensitivity. This is in agreement with previously published test results (ref. 2).

Thus, the results indicate that hardness testing is a useful method for determining the likelihood of whether heat treatments will be susceptible to time-dependent notch sensitive behavior.

SUMMARY OF RESULTS

A study was made of the time-dependent edge-notch sensitivity of 0.030 inch (.75mm) Inconel 718 sheet. Consideration of the results led to the following:

(1) Time-dependent notch sensitivity was observed at temperatures from 900 ° to 1200°F (482 - 649°C). No reasons were evident why similar behavior could not be expected at prolonged times at lower temperatures. Notched to smooth rupture strength ratios fell to as low as 0.37. At 1400°F (760°C) ratios of about 1.0 were obtained, i.e. no notch sensitivity was observed.

(2) Both smooth and notched creep-rupture specimens failed by relatively slow intergranular crack initiation and growth followed by transgranular fracture. Time-dependent notch sensitivity was due to premature initiation of intergranular cracks. This was caused by localized creep deformation resulting from relaxation of the high stresses introduced by the presence of the edge-notches.

Necessary conditions for time-dependent notch sensitivity were (i) the notched specimen loads had to be below the approximate 0.2 percent smooth specimen offset yield strength; and (ii) test data from smooth specimens had to indicate that small amounts of creep used up large fractions of creep-rupture life. These observations are identical to those reported for Waspaloy (ref. 1).

(3) The severity of the time-dependent notch sensitivity was dependent on the heat treatment. Decreasing the solution temperature or increasing time and/or temperature of the aging treatment decreased the susceptibility to the time-dependent notch sensitivity. No time-dependent notch sensitivity was observed for materials heat treated 10 hours at 1700°F (927°C) plus 48 hours at 1350°F (732°C) and 1 hour at 1700°F (927°C) plus 2 hours at 1550°F (843°C).

It should be noted that the commonly used aging treatments: 1350°F (732°C)/8 hours, F.C. to 1200°F (649°C) in 12 hours, A.C. {after solution treatment at 1950°F (1066°C)} and 1325°F (718°C)/ 8 hours, F.C. to 1150°F (621°C) in 10 hours, A.C. {after solution treatment at 1750°F (954°C)} resulted in notch sensitive behavior.

(4) The dislocation motion mechanism varied with heat treatment and test conditions. Dislocations sheared the bct Ni_3Cb particles (and gamma prime) smaller than a "critical size". Larger particles were by-passed by dislocations. These gave rise to localized and homogeneous deformation respectively. Localized deformation increased the susceptibility to notch sensitivity. In consequence, a correlation existed between the occurrence of time-dependent notch sensitivity and the nature of the dislocation motion mechanism operative.

Results indicated that room temperature hardness tests can be used to indicate particle size relative to the "critical".

Similar relations were evident from the study of Waspaloy which differs in that it age hardens solely due to gamma prime, $Ni_3(Ti, Al)$ (ref. 1). For Inconel 718, although gamma prime is present, the principal strengthening phase is bct Ni_3Cb.

REFERENCES

1. Wilson, D. J. and Freeman, J. W.: "Sensitivity of the Creep-Rupture Properties of Waspaloy Sheet to Sharp-Edged Notches in the Temperature Range of 1000° to 1400°F". Prepared under Grant No. NGL-23-005-005 (NASA CR-1849) for NASA by the University of Michigan, Ann Arbor, June 1971.

2. Cullen, T. M. and Freeman, J. W.: "The Mechanical Properties of Inconel 718 Sheet Alloy at 800°F, 1000°F and 1200°F". Prepared under Grant No. NsG-124-61 (NASA CR-268) for NASA by the University of Michigan, Ann Arbor, July, 1965.

3. Eiselstein, H. L.: "Metallurgy of a Columbium-Hardened Nickel-Chromium-Iron Alloy". ASTM-STP No. 369, p. 62, 1965.

4. Bigelow, W. C., Amy, J. A., and Brockway, L. O.: "Electron Microscope Identification of the Gamma Prime Phase of Nickel-Base Alloys". Proc. ASTM, Vol. 56, p. 945, 1956.

5. Paulonis, D. F., Oblak, J. M., and Duvall, D. S.: "Precipitation in Nickel-Base Alloy 718". Trans. ASM, Vol. 62, p. 611, 1969.

6. Kirman, I. and Warrington, D. H.: "Identification of the Strengthening Phase in Fe-Ni-Cr-Nb Alloys". JISI, Vol. 205, p. 1264, 1967.

7. Kirman I. and Warrington, D. H.: "The Precipitation of Ni_3Nb Phases in a Ni-Fe-Cr-Nb Alloy", Metall. Trans., Vol. 1, p. 2667, 1970.

8. Kotval, P. S.: "The Microstructure of Superalloys". Metallography, Vol. 1, p. 251, 1969.

9. Decker, R. F.: "Strengthening Mechanisms in Nickel-Base Superalloys". Presented at the Steel Strengthening Mechanisms Symposium, Zurich, Switzerland, May, 1969.

TABLE 1

TENSILE PROPERTIES OF 0.030 INCH (.75mm) THICK INCONEL 718 SHEET AT 1000° AND 1200°F (538° AND 649°C)

Heat Treatment	Test Temperature (°F)	Test Temperature (°C)	Smooth Specimen Properties: Tensile Strength (ksi)	Tensile Strength MN/m^2	0.2% Offset Yield Strength (ksi)	0.2% Offset Yield Strength MN/m^2	Y.S./T.S.	Elong. (%)	R.A. (%)	Notched Tensile Strength (ksi)	Notched Tensile Strength MN/m^2	N/S Tensile Strength Ratio
10 hrs. at 1950°F (1066°C)	1000	538	145.4	1002	118.5	817	0.82	15.4	23	149.0	1027	1.02
+ 48 hrs. at 1350°F (732°C)	1200	649	142.4	982	116.5	803	0.82	8.1	14	135.3	933	0.95
1 hr. at 1950°F (1066°C)	1000	538	160.1	1104	129.5	893	0.81	22.5	30	152.4	1051	0.95
+ 48 hrs. at 1350°F (732°C)	1200	649	157.5	1086	130.5	900	0.83	8.4	14	147.5	1017	0.94
1 hr. at 1950°F (1066°C)	1000	538	134.9	930	90.5	624	0.67	27.0	34	113.6	783	0.84
+ 2 hrs. at 1550°F (843°C)	1200	649	133.4	920	99.5	686	0.75	11.3	18	116.0	800	0.87
1 hr. at 1950°F (1066°C)	1000	538	134.2	925				32.8	30	102.1	704	0.76
+ 24 hrs. at 1550°F (843°C)	1200	649	132.6	914	83.5	576	0.63	14.6	17	108.0	745	0.81
10 hrs. at 1800°F (982°C)	1000	538	162.0	1117	122.0	841	0.75	19.6	23	135.9	937	0.84
+ 48 hrs. at 1350°F (732°C)	1200	649	147.6	1018	120.5	831	0.82	16.7	15	140.8	971	0.95
10 hrs. at 1700°F (927°C)	1000	538	157.5	1086	116.0	800	0.75	15.5	24			
+ 3 hrs. at 1325°F (718°C)												
10 hrs. at 1700°F (927°C)	1000	538	166.0	1144	116.5	803	0.71	15.9	23	125.1	863	0.75
+ 48 hrs. at 1350°F (732°C)	1200	649	143.7	991	115.5	796	0.80	16.8	26	138.1	952	0.96
1 hr. at 1700°F (927°C)	1000	538	165.2	1139	135.0	931	0.82	12.9	27	154.9	1068	0.94
+ 3 hrs. at 1325°F (718°C)	1200	649	160.2	1105	136.0	938	0.85	8.3	22	140.3	967	0.88
1 hr. at 1700°F (927°C)	1000	538	149.0	1027	101.5	700	0.68	19.8	27	121.9	841	0.82
+ 2 hrs. at 1550°F (843°C)	1200	649	135.4	934	103.0	710	0.76	14.2	20	121.2	836	0.89

TABLE 2

SMOOTH AND NOTCHED (Kt>20) SPECIMEN TENSILE AND CREEP-RUPTURE PROPERTIES AT 900°F TO 1400°F (482-760°C) FOR 0.030-INCH (.75mm) THICK INCONEL 718 SHEET HEAT TREATED 10 HOURS AT 1950°F (1066°C) PLUS 48 HOURS AT 1350°F (732°C)

SMOOTH SPECIMENS

Test Temperature (°F)	Test Temperature (°C)	Stress (ksi)	Stress (MN/m^2)	Rupture Time (hrs.)	Larson-Miller Parameter (C=20) x 10^{-3}	Elong. (%)	R.A. (%)	Min. Creep Rate (%/ hr.)	Intergranular Crack Length (%)
1000	538	145.4	1002	Tensile		15.4	23		0
		130	896	21.5	31.15	7.4	13		8
		115	793	333.7	32.88	3.4	9	0.00033	11
		105	724	3332.2ph	34.34			-ve.	
1100	593	100	690	431.3	35.31	1.5	4	0.00041	18
		85	586	9691.4	37.42	1.0	3	-ve.	26
1200	649	142.4	982	Tensile		8.1	14		0
		90	621	216.1	37.07	1.1	8	0.00067	26
		80	552	282.4ph	37.27			0.00055	
		70	483	924.8	38.12	1.1	5	0.00014	34
		60	414	3564.0	39.10	1.3	4	0.000038	44
1400	760	30	207	292.1	41.79	1.9	3	0.00091	60

ph=failed at pin hole

NOTCHED SPECIMENS

Test Temperature (°F)	Test Temperature (°C)	Stress (ksi)	Stress (MN/m^2)	Rupture Time (hrs.)	Larson-Miller Parameter (C=20) x 10^{-3}	Intergranular Crack Length (%)	N/S Strength Ratio
900	482	120	827	360.0	30.68	16	
		90	621	1075.1	31.32	27	
1000	538	149.0	1027	Tensile		0	1.02
		120	827	18.7	31.06	6	0.92
		90	621	100.7	32.12	26	0.74
		60	414	682.7	33.34	54	0.53
		50	345	2342.4	34.12	38	0.47
1100	593	80	552	31.6	33.54	35	0.71
		60	414	212.8	34.83		0.58
		50	345	2121.8	36.39	39	0.55
1200	649	135.3	933	Tensile		1	0.95
		60	414	3.1	34.02	32	0.46
		50	345	1.5	33.49	52	0.37
		40	276	5774.6	39.44	73	0.72
1400	760	30	207	451.8	42.14	79	1.13

TABLE 3

SMOOTH AND NOTCHED (Kt>20) SPECIMEN TENSILE AND CREEP-RUPTURE PROPERTIES AT 900°F TO 1400°F (482-760°C) FOR 0.030-INCH (.75mm) THICK INCONEL 718 SHEET HEAT TREATED 1 HOUR AT 1950°F (1066°C) PLUS 48 HOURS AT 1350°F (732°C)

SMOOTH SPECIMENS

Test Temperature (°F)	Test Temperature (°C)	Stress (ksi)	Stress MN/m^2	Rupture Time (hrs.)	Larson-Miller Parameter (C=20) x 10^{-3}	Elong. (%)	R. A. (%)	Min. Creep Rate (% / hr.)	Intergranular Crack Length (%)
1000	538	160.1	1104	Tensile		22.5	30		0
		140	965	385.0	32.97	2.7	11	0.00043	10
		130	896	2506.1	34.16	1.6	7	0.000022	14
1100	593	120	827	1.4	31.43	4.2	10		8
		110	758	1930.1	36.33	1.3	7	0.00015	22
		100	690	6833.6	37.18	1.6	6	0.000013	23
1200	649	157.5	1086	Tensile		8.4	14		0
		90	621	707.6	37.93	1.3	8	0.00028	24
		80	552	228.9ph	37.12			0.00025	
		70	483	2469.5	38.83	0.8	2	0.000073	40
		60	414	6261.4	39.50	0.8	2	-ve.	40
		60	414	8168.8	39.69		2	0.000026	48
1400	760	30	207	384.0	42.01	2.1	5	0.00077	24

NOTCHED SPECIMENS

Test Temperature (°F)	Test Temperature (°C)	Stress (ksi)	Stress MN/m^2	Rupture Time (hrs.)	Larson-Miller Parameter (C=20) x 10^{-3}	Intergranular Crack Length (%)	N/S Strength Ratio
900	482	130	896	34.6	29.29	6	
		90	621	1745.1	32.21	41	
1000	538	152.4	1051	Tensile		0	0.95
		90	621	86.2	32.03	29	0.60
		70	483	221.2	32.62	41	0.49
		60	414	9262 Discontinued			>0.49
1100	593	80	552	10.8	32.81	30	0.55
		60	414	184.1	34.73	44	0.47
		50	345	8902 Discontinued			>0.51
1200	649	147.5	1017	Tensile		1	0.94
		50	345	46.3	35.96	45	0.41
		45	310	330.2	37.38	70	0.45
1400	760	30	207	419.8	42.08	75	1.02

ph=failed at pin hole

TABLE 4

SMOOTH AND NOTCHED (Kt>20) SPECIMEN TENSILE AND CREEP-RUPTURE PROPERTIES AT 900°F TO 1400°F (482-760°C) FOR 0.030-INCH (.75mm) THICK INCONEL 718 SHEET HEAT TREATED 1 HOUR AT 1950°F (1066°C) PLUS 2 HOURS AT 1550°F(843°C)

SMOOTH SPECIMENS

Test Temperature (°F)	Test Temperature (°C)	Stress (ksi)	Stress MN/m^2	Rupture Time (hrs.)	Larson-Miller Parameter (C=20) x 10^{-3}	Elong. (%)	R.A. (%)	Min. Creep Rate (%/hr.)	Intergranular Crack Length (%)
1000	538	134.9	930	Tensile		27.0	34		0
		130	896	63.5	31.83	18.7	19		6
		120	827	154.1	32.39	11.9	14	0.00020	11
		115	793	4402.3	34.52	7.3	13	0.000005	17
1100	593	110	758	138.2	34.54	7.8	13	0.00030	20
		100	690	385.2ph	35.23			<0.00047	
		85	586	17900 In Progress					
1200	649	133.4	92[illegible]	Tensile		11.3	18		1
		100	690	72.6	36.29	3.2	9	0.0025	26
		80	552	938.6	38.13	1.8	6	0.00021	39
		70	483	2094.6	38.71	1.6	5	0.00011	44
1400	760	60	414	16.9	39.48	3.5	8		43
		30	207	508.0	42.23	2.9	5	0.00091	98

ph = failed at pin hole

NOTCHED SPECIMENS

Test Temperature (°F)	Test Temperature (°C)	Stress (ksi)	Stress MN/m^2	Rupture Time (hrs.)	Larson-Miller Parameter (C=20) x 10^{-3}	Intergranular Crack Length (%)	N/S Strength Ratio
900	482	80	552	5763.6	32.31	30	
1000	538	113.6	783	Tensile		0	0.84
		90	621	173.8	32.47	29	0.72
		75	517	752.9	33.40	34	0.62
		65	448	1138.9	33.66	49	0.55
1100	593	80	552	20.0	33.23	26	0.67
		60	414	704.8	35.6[illegible]	59	0.61
1200	649	116.0	800	Tensile		3	0.87
		80	552	4.1	34.22	29	0.66
		60	414	225.0	37.10	44	0.65
		50	345	5003.7	39.34	51	0.79
1400	760	50	345	2.8	38.03	44	0.58
		40	276	148.3	41.24	59	1.05
		30	207	408.5	42.06	71	0.95
		30	207	493.8	42.21	73	1.00

TABLE 5

SMOOTH AND NOTCHED (Kt>20) SPECIMEN TENSILE AND CREEP-RUPTURE PROPERTIES AT 900°F TO 1400°F (482-760°C) FOR 0.030-INCH (.75mm) THICK INCONEL 718 SHEET HEAT TREATED 1 HOUR AT 1950°F (1066°C) PLUS 24 HOURS AT 1550°F (843°C)

SMOOTH SPECIMENS

Test Temperature (°F)	Test Temperature (°C)	Stress (ksi)	Stress (MN/m^2)	Rupture Time (hrs.)	Larson-Miller Parameter (C=20) x 10^{-3}	Elong. (%)	R.A. (%)	Min. Creep Rate (% / hr.)	Intergranular Crack Length (%)
1000	538	134.2	925	Tensile		32.8	30		0
		125	862	40.9	31.55	24.9	23	0.090	3
		115	793	1856.7	33.97	10.2	12	0.00014	10
1100	593	110	758	90.5	34.25	9.9	12	0.0034	15
		100	690	3528.1	36.73	4.7	6	0.000005	20
1200	649	132.6	914	Tensile		14.6	17		1
		100	690	11.6	34.97	6.9	13	0.17	19
		80	552	926.5	38.12	3.2	6	0.00037	23
		70	483	2480.9	38.84	2.0	5	0.000080	32
1400	760	30	207	558.1	42.31	8.7	12	0.0011	92

NOTCHED SPECIMENS

Test Temperature (°F)	Test Temperature (°C)	Stress (ksi)	Stress (MN/m^2)	Rupture Time (hrs.)	Larson-Miller Parameter (C=20) x 10^{-3}	Intergranular Crack Length (%)	N/S Strength Ratio
1000	538	102.1	704	Tensile		0	0.76
		90	621	253.5	32.71	14	0.76
		80	552	6862 Discontinued			>0.73
		60	414	11000 Discontinued			>0.55
1100	593	85	586	40.6	33.71	23	0.70
		75	517	788.5	35.72	43	0.74
		60	414	9044 Discontinued			>0.61
1200	649	108.0	745	Tensile		4	0.81
		60	414	18.9	35.32	26	0.61
		60	414	40.3	35.86	30	0.63
		50	345	5588.8	39.42	48	0.81
1400	760	30	207	373.8	41.98	53	0.88

TABLE 6

SMOOTH AND NOTCHED (Kt>20) SPECIMEN TENSILE AND CREEP-RUPTURE PROPERTIES AT 900° TO 1400°F (482 - 760°C) FOR 0.030-INCH (.75mm) THICK INCONEL 718 SHEET HEAT TREATED 10 HOURS AT 1800°F (982°C) PLUS 48 HOURS AT 1350°F (732°C)

SMOOTH SPECIMENS

Test Temperature (°F)	Test Temperature (°C)	Stress (ksi)	Stress MN/m^2	Rupture Time (hrs.)	Larson-Miller Parameter (C=20) x 10^{-3}	Elong. (%)	R. A. (%)	Min. Creep Rate (%/hrs.)	Intergranular Crack Length (%)
1000	538	162.0	1117	Tensile		19.6	23		0
		130	896	219.7	32.62	7.0	10	0.0175	9
		115	793	13437.5	35.23	3.1	6	0.000046	13
1100	593	115	793	336.9	35.14	4.5	10	0.0052	12
		100	690	1266.5	36.04	3.0	8	0.00035	19
		90	621	5793.2	37.07	3.0	10	0.00010	26
1200	649	147.6	1018	Tensile		16.7	15		1
		100	690	34.0	35.74	7.2	12	0.095	5
		80	552	410.4	37.54	6.2	11	0.027	26
		70	483	1420.9	38.43	6.0	13	0.00070	42
1400	760	30	207	231.8	41.60	17.4	42	0.0056	52

NOTCHED SPECIMENS

Test Temperature (°F)	Test Temperature (°C)	Stress (ksi)	Stress MN/m^2	Rupture Time (hrs.)	Larson-Miller Parameter (C=20) x 10^{-3}	Intergranular Crack Length (%)	N/S Strength Ratio
900	482	90	621	980.3	31.27	17	
1000	538	135.9	937	Tensile		0	0.84
		90	621	131.4	32.29	20	0.68
		75	517	238.4	32.67	26	0.58
		65	448	305.8	32.83	30	0.51
1100	593	90	621	8.2	32.63	19	0.67
		75	517	105.2	34.35	31	0.64
		60	414	12867 Discontinued			>0.68
1200	649	140.8	971	Tensile		0	0.95
		90	621	14.0	35.10	32	0.85
		75	517	5.1	34.37	39	0.68
		60	414	1406.5	38.42	58	0.87
1400	760	30	207	215.4	41.54	70	0.99

TABLE 7

SMOOTH AND NOTCHED (Kt>20) SPECIMEN TENSILE AND CREEP-RUPTURE PROPERTIES AT 900°F TO 1400°F (482-760°C) FOR 0.030-INCH (.75mm) THICK INCONEL 718 SHEET HEAT TREATED 10 HOURS AT 1700°F (927°C) PLUS 3 HOURS AT 1325°F (718°C)

SMOOTH SPECIMENS

Test Temperature (°F)	Test Temperature (°C)	Stress (ksi)	Stress MN/m^2	Rupture Time (hrs.)	Larson-Miller Parameter (C=20) x 10^{-3}	Elong. (%)	R.A. (%)	Min. Creep Rate (%/hr.)	Intergranular Crack Length (%)
1000	538	157.5	1086	Tensile		15.5	24		0
		145	1000	26.5	31.28	21.6	22		1
		130	896	391.1	32.98	9.0	11	0.0048	5
1100	593	120	827	95.0	34.38	8.4	13	0.022	8
		90	621	2138.2	36.39	7.8	13	0.00024	11
1200	649	85	586	79.4	36.35	6.0	18	0.012	65
		60	414	1061.8	38.22	12.6	16	0.0016	
		50	345	3766.9	39.14	15.8	16	0.00038	
1400	760	20	137	506.9	42.23	32.3	50	0.0068	

NOTCHED SPECIMENS

Test Temperature (°F)	Test Temperature (°C)	Stress (ksi)	Stress MN/m^2	Rupture Time (hrs.)	Larson-Miller Parameter (C=20) x 10^{-3}	Intergranular Crack Length (%)	N/S Strength Ratio
900	482	90	621	2639.1	31.85	12	
1000	538	90	621	598.6	33.25	26	0.72
		75	517	7680 Discontinued			>0.95
1100	593	90	621	77.4	34.15	25	0.75
		80	552	951.2	35.85		0.83
1200	649	85	586	45.2	35.95	31	0.91
		60	414	1024.2	38.20	55	1.00
1400	760	30	207	103.9	40.95	70	1.00

TABLE 8

SMOOTH AND NOTCHED (Kt>20) SPECIMEN TENSILE AND CREEP-RUPTURE PROPERTIES AT 900° TO 1400°F (482-760°C) FOR 0.030-INCH (.75mm) THICK INCONEL 718 SHEET HEAT TREATED 10 HOURS AT 1700°F (927°C) PLUS 48 HOURS AT 1350°F (732°C)

SMOOTH SPECIMENS

Test Temperature (°F)	Test Temperature (°C)	Stress (ksi)	Stress MN/m^2	Rupture Time (hrs.)	Larson-Miller Parameter (C=20) x 10^{-3}	Elong. (%)	R. A. (%)	Min. Creep Rate (%/hr.)	Intergranular Crack Length (%)
1000	538	166.0	1144	Tensile		15.9	23		0
		130	896	125.7	32.26	11.7	14	0.050	2
		120	827	1382.8	33.78	5.6	8	0.0022	5
		115	793	1163.2	33.68	6.1	10	0.0023	6
1100	593	100	690	192.2	34.76	11.7	14	0.036	10
		85	586	1574.0	36.19	2.7	6	0.00073	22
1200	649	143.7	991	Tensile		16.8	26		0
		85	586	42.5	35.90	9.2	20	0.087	15
		60	414	530.7	37.72	12.0	18	0.0045	30
		50	345	3378.9	39.06	12.4	21	0.00072	47
1400	760	30	207	96.9	40.89	18.0	43	0.040	32
		20	137	439.0	42.11	32.7	45	0.013	

ph = failed at pin hole

NOTCHED SPECIMENS

Test Temperature (°F)	Test Temperature (°C)	Stress (ksi)	Stress MN/m^2	Rupture Time (hrs.)	Larson-Miller Parameter (C=20) x 10^{-3}	Intergranular Crack Length (%)	N/S Strength Ratio
900	482	90	621	13140 Discontinued			
1000	538	125.1	863	Tensile		0	0.75
		90	621	299.2	32.81	20	0.72
		75	517	7656 Discontinued			>0.70
1100	593	90	621	167.8	34.67	28	0.89
		80	552	847.9	35.77	33	0.90
1200	649	138.1	952	Tensile		1	0.96
		80	552	53.2	36.06	40	0.98
		60	414	468.2	37.63	52	0.94
		50	345	2422.0	38.82	65	0.97
1400	760	35	241	57.0ph	40.47	54	1.00
		30	207	103.0	40.94	57	1.00
		20	137	483.0	42.19	67	1.02

TABLE 9

SMOOTH AND NOTCHED (Kt>20) SPECIMEN TENSILE AND CREEP-RUPTURE PROPERTIES AT 900°F TO 1400°F (482-760°C) FOR 0.030-INCH (.75mm) THICK INCONEL 718 SHEET HEAT TREATED 1 HOUR AT 1700°F (927°C) PLUS 3 HOURS AT 1325°F (718°C)

SMOOTH SPECIMENS

Test Temperature (°F)	Test Temperature (°C)	Stress (ksi)	Stress MN/m^2	Rupture Time (hrs.)	Larson-Miller Parameter (C=20) x 10^{-3}	Elong. (%)	R.A. (%)	Min. Creep Rate (%/hr.)	Intergranular Crack Length (%)
1000	538	165.2	1139	Tensile		12.9	27		0
		145	1000	168.9	32.45	12.1	14	0.0048	2
		130	896	5613.4	34.67	3.5	12	0.000086	13
1100	593	125	862	184.8	34.74	1.65	11	0.001	12
		115	793	406.6	35.27	1.60	6	0.00072	13
		100	690	2727.0	36.56	3.4	11	0.00017	26
1200	649	160.2	1105	Tensile		8.3	22		0
		100	690	53.0	36.06	1.75	9	0.0096	24
		85	586	291.1	37.29		23		18
		85	586	117.1	36.63	5.0	15	0.0061	23
		65	448	937.4	38.13	3.0	7	0.00096	36
		65	448	1501.2	38.47		9		32
1400	760	30	207	144.0	41.21	10.8	27	0.013	60

NOTCHED SPECIMENS

Test Temperature (°F)	Test Temperature (°C)	Stress (ksi)	Stress MN/m^2	Rupture Time (hrs.)	Larson-Miller Parameter (C=20) x 10^{-3}	Intergranular Crack Length (%)	N/S Strength Ratio
900	482	90	621	2547.2	31.83	30	
1000	538	154.9	1068	Tensile		0	0.94
		130	896	17.5	31.01	7	0.86
		90	621	152.6	32.39	23	0.62
		75	517	239.4	32.67	39	0.53
		65	448	1561.3	33.86	51	0.49
1100	593	90	621	30.7	33.52	28	0.60
		80	552	131.4	34.50	26	0.59
1200	649	140.3	967	Tensile		1	0.88
		85	586	2.6	33.89	35	0.63
		75	517	109.8	36.59	37	0.83
		65	448	1031.9	38.20	52	0.99
1400	760	30	207	105.9	40.97	61	0.94

TABLE 10

SMOOTH AND NOTCHED (Kt>20) SPECIMEN TENSILE AND CREEP-RUPTURE PROPERTIES AT 900°F TO 1400°F (482-760°C) FOR 0.030-INCH (.75mm) THICK INCONEL 718 SHEET HEAT TREATED 1 HOUR AT 1700°F (927°C) PLUS 2 HOURS AT 1550°F (843°C)

SMOOTH SPECIMENS

Test Temperature (°F)	Test Temperature (°C)	Stress (ksi)	Stress (MN/m^2)	Rupture Time (hrs.)	Larson-Miller Parameter (C=20) x 10^{3}	Elong. (%)	R. A. (%)	Min. Creep Rate (%/ hr.)	Intergranular Crack Length (%)
1000	538	149.0	1027	Tensile		19.8	27		0
		145	1000	2.0	29.64	24.3	32	1.7	1
		130	896	18.1	31.04	19.4	25		1
		120	827	785.0	33.43	9.3	12	0.0016	7
		115	793	4385.2	34.52	2.6	6	0.000082	7
1100	593	110	758	486.3	35.39	5.8	9	0.0018	9
		100	690	703.6	35.64	1.5	7	0.00056	19
		85	586	4078.6	36.83	2.2	4	0.00010	23
1200	649	135.4	934	Tensile		14.2	20		1
		100	690	21.9	35.42	9.3	15		12
		85	586	154.3	36.83	4.0	9	0.0085	20
		65	448	1011.8	38.19	6.2	7	0.0013	20
1400	760	30	207	160.1	41.30	21.4	34	0.030	33
		20	137	535.2	42.28	25.4	35	0.0053	

NOTCHED SPECIMENS

Test Temperature (°F)	Test Temperature (°C)	Stress (ksi)	Stress (MN/m^2)	Rupture Time (hrs.)	Larson-Miller Parameter (C=20) x 10^{3}	Intergranular Crack Length (%)	N/S Strength Ratio
1000	538	121.9	840.5	Tensile		0	0.82
		90	621	3205.7	34.32	23	0.78
		75	517	5133 Discontinued			>0.66
1100	593	90	621	748.4	35.68	28	0.92
1200	649	121.2	836	Tensile		0	0.89
		65	448	223.3	37.10	42	0 80
		50	345	1677	38.57	59	0.82
1400	760	30	207	92.5	40.86	61	0.85

TABLE 11

X-RAY DIFFRACTION DATA OF EXTRACTED RESIDUES OF INCONEL 718 IN THE HEAT TREATED CONDITIONS

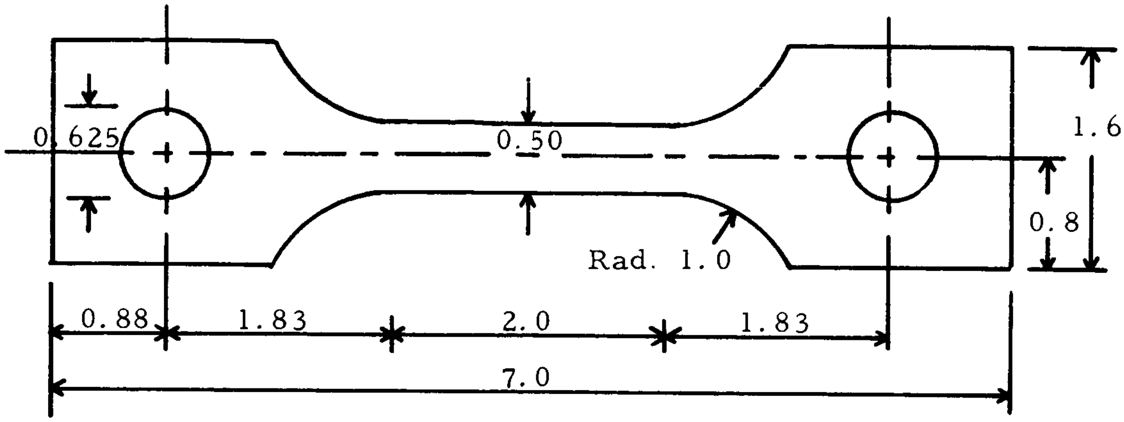
Solution Treatment	10 hrs. at 1950°F (1066°C)	1 hr. at 1950°F (1066°C)	1 hr. at 1950°F (1066°C)	1 hr. at 1950°F (1066°C)	10 hrs. at 1800°F (928°C)	10 hrs. at 1700°F (927°C)	10 hrs. at 1700°F (927°C)	1 hr. at 1700°F (927°C)	1 hr. at 1700°F (927°C)	
Aging Treatment	48 hrs. at 1350°F (732°C)	48 hrs. at 1350°F (732°C)	2 hrs. at 1550°F (843°C)	24 hrs. at 1550°F (843°C)	48 hrs. at 1350°F (732°C)	3 hrs. at 1325°F (718°C)	48 hrs. at 1350°F (732°C)	3 hrs. at 1325°F (718°C)	2 hrs. at 1550°F (843°C)	Indicated Phases
	d(A) I	d(A) I	d(A) I	d(A) I	d(A) I	d(A) I	d(A) I	d(A) I	d(A) I	
	3.25 vw	3.26 vvw	3.25 vw	3.25 w	3.25 vw	3.26 vvw	3.26 vvw		3.25 w	γ''
					2.643vvw	2.646w	2.659vvw		2.644w	?
	2.552w	2.57 vw	2.564w	2.561w	2.555w		2.567vvw	2.552w	2.553w	MC
					2.385vvw	2.394vvw		2.388vvw		L?
								2.330w		?
			2.66vvw		2.257vw	2.272w	2.272vw		2.264vvw	β
						2.231m	2.232m	2.232m		β
	2.214w	2.226vw	2.223w	2.217w	2.219w		2.222vw		2.221w	MC, L?
	2.116to*	2.123to*	2.111vs	2.108vs	2.104vs	2.117s	2.116s	2.118m	2.108vs	β
	2.078vs	2.087vs	2.080vw	2.080m	2.083vs	2.08 vw	2.081s	2.080w	2.079s	γ'/γ''
				2.040vvw	2.033vw	2.033vvw		2.043w	2.038vw	L?
			1.994w	1.992vw	1.992m	2.000m	2.000s	2.003m	1.996m	β, L?
			1.969w	1.948w	1.965s	1.977s	1.973s	1.975s	1.968s	β
	1.850vw		1.854vw	1.846w					1.855w	γ'/γ''
	1.809s	1.813m	1.818w	1.808s	1.805s		1.807m	1.807vvw	1.809m	γ'/γ''
	1.568w	1.575vvw	1.569vw	1.567w	1.567vw	1.573vvw	1.573vvw	1.574w	1.566vw	MC
					1.545vvw	1.548vvw	1.549vw		1.546vvw	?
			1.535vw		1.530w	1.536m	1.534w	1.536w	1.535w	?
	1.337w	1.339vw	1.338vw	1.337w	1.338w		1.337vvw	1.344vvw	1.338w	MC
	1.292w	1.296vw	1.298vw	1.294m	1.300w	1.302w	1.302w	1.303vvw	1.300w	β, γ'/γ''
	1.278m	1.280w	1.280vw	1.280m	1.278m	1.281w	1.280w	1.281vvw	1.280w	MC, ?
					1.195w	1.195w		1.199w		L?
			1.193vw	1.188vvw	1.189w	1.191w	1.193w	1.193vvw	1.192w	?
	1.110vw	1.110w	1.107vw	1.110m	1.106w	1.108w	1.109w	1.111w	1.108w	β, ?
Phases										
MC	w	w	w	w	w	w	w	w	w	
γ'/γ''	s	s	s	vs	s	w	m	w	s	
β	0	0	w	w	m	s	s	s	m	

*Values at extremes of wide reflection

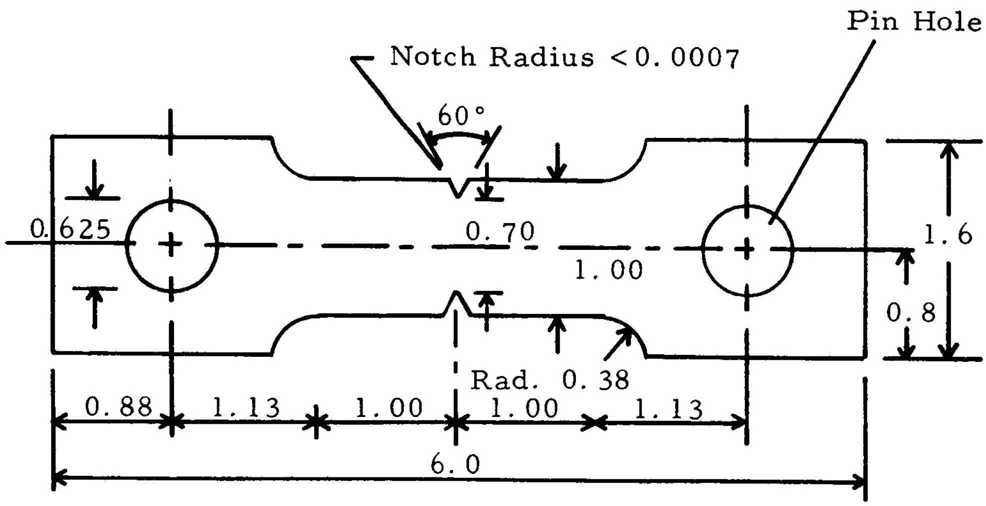
Intensities I:s = Strong, m = Medium, w = Weak, v = Very

Phases: MC = Cb, Ti(C, N), $\gamma' = Ni_3(Al, Ti)$, $\gamma'' = Ni_3Cb$(bct), $\beta = Ni_3Cb$(orthorhombic)

L = "Laves", $Fe_2(Ti, Cb)$



Smooth (unnotched) Specimen ($K_t = 1.0$)



Sharp Edge Notched Specimen ($K_t > 20$)

Figure 1. Types of test specimens (all dimensions in inches).

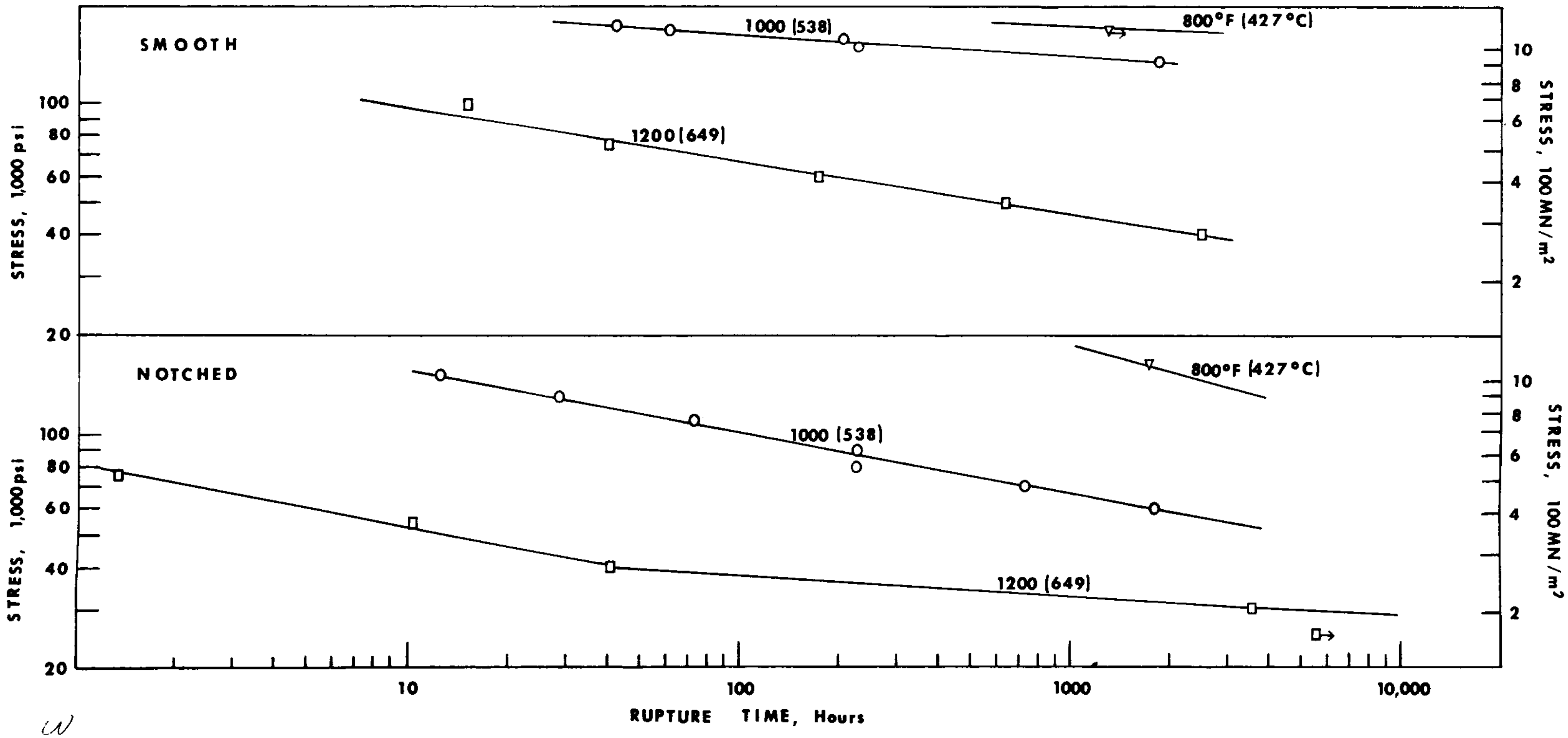


Figure 2. Stress versus rupture time data at temperatures from 800°F to 1200°F (427 - 649°C) obtained from smooth and notched specimens of Inconel 718 sheet cold worked 20 percent and aged (ref. 2). Time-dependent notch sensitivity was evident at 1000°F (538°C) and 1200°F (649°C).

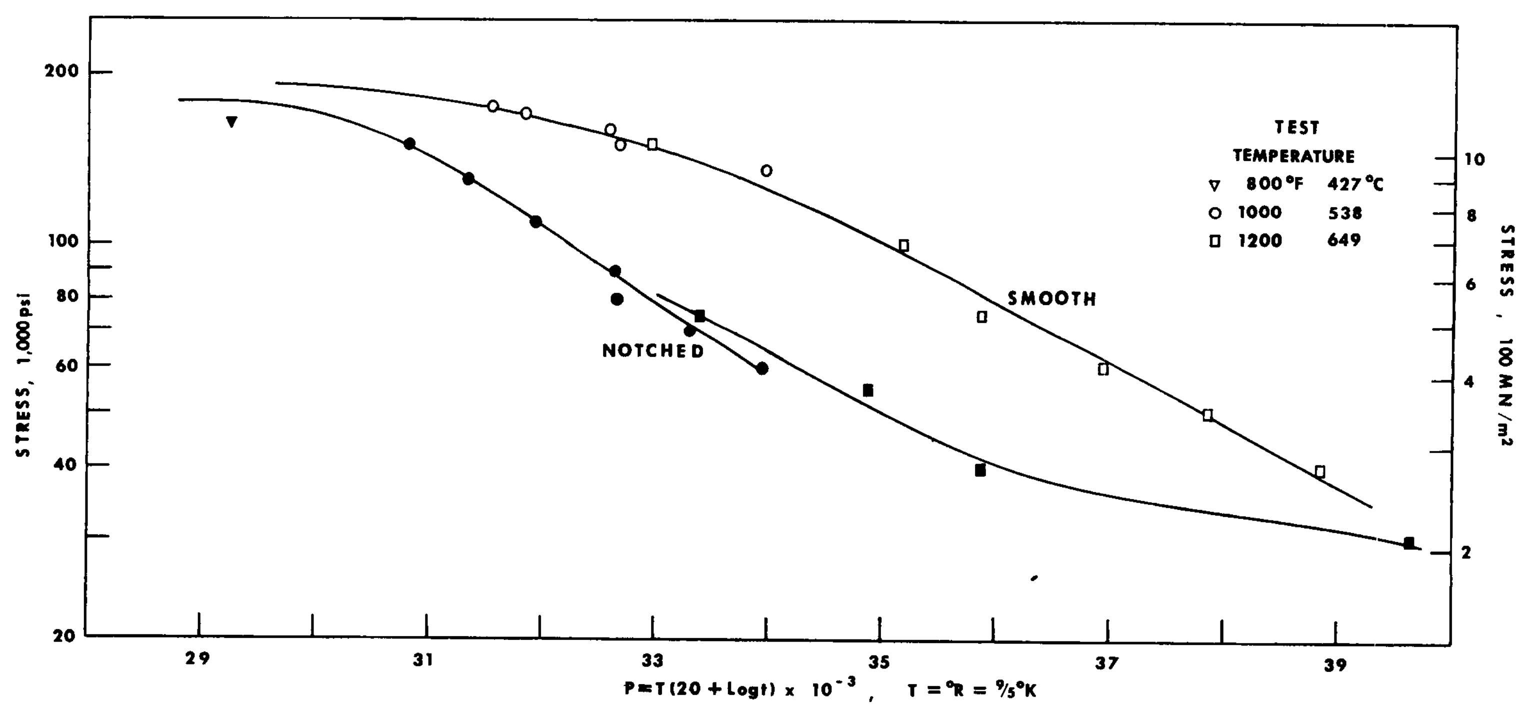


Figure 3. Time-temperature dependence of the rupture strengths of smooth and notched specimens of Inconel 718 cold worked and 20 percent and aged.

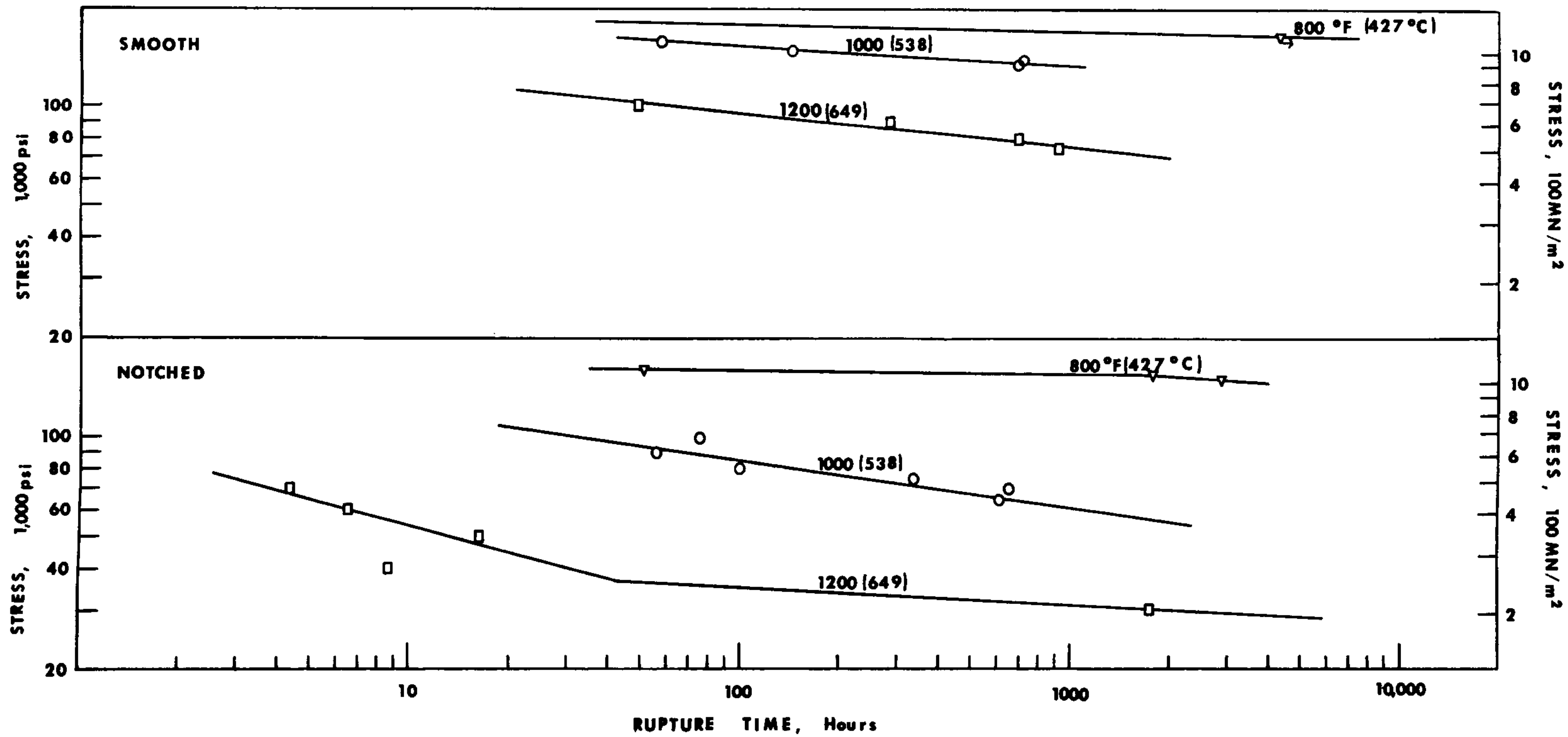


Figure 4. Stress versus rupture time data at temperatures from 800°F to 1200°F (427 - 649°C) obtained from smooth and notched specimens of Inconel 718 sheet heat treated I hour at 1950°F (1066°C) and aged (ref. 2). Time-dependent notch sensitivity occurred for tests at 1000°F (538°C) and 1200°F (649°C).

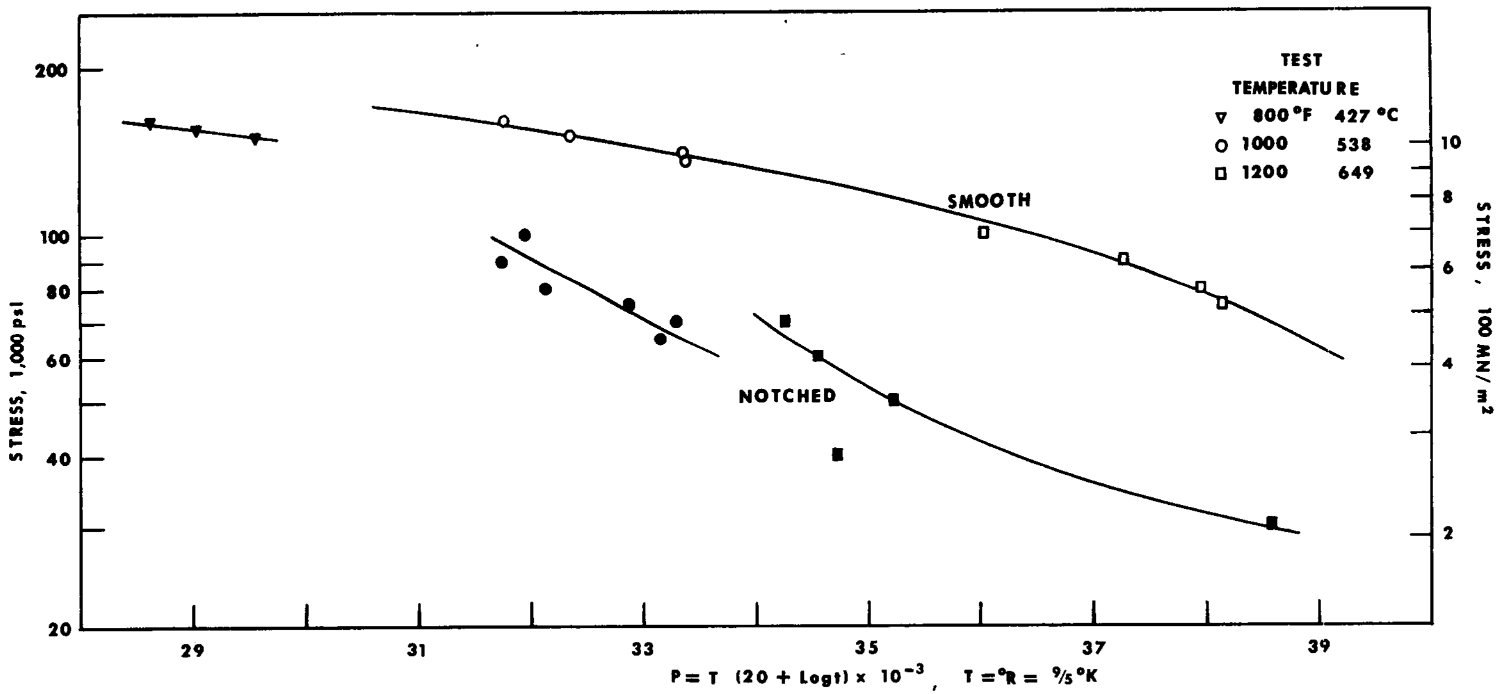


Figure 5. Time-temperature dependence of the rupture strengths of smooth and notched specimens of Inconel 718 sheet heat treated 1 hour at 1950°F (1066°C) and aged.

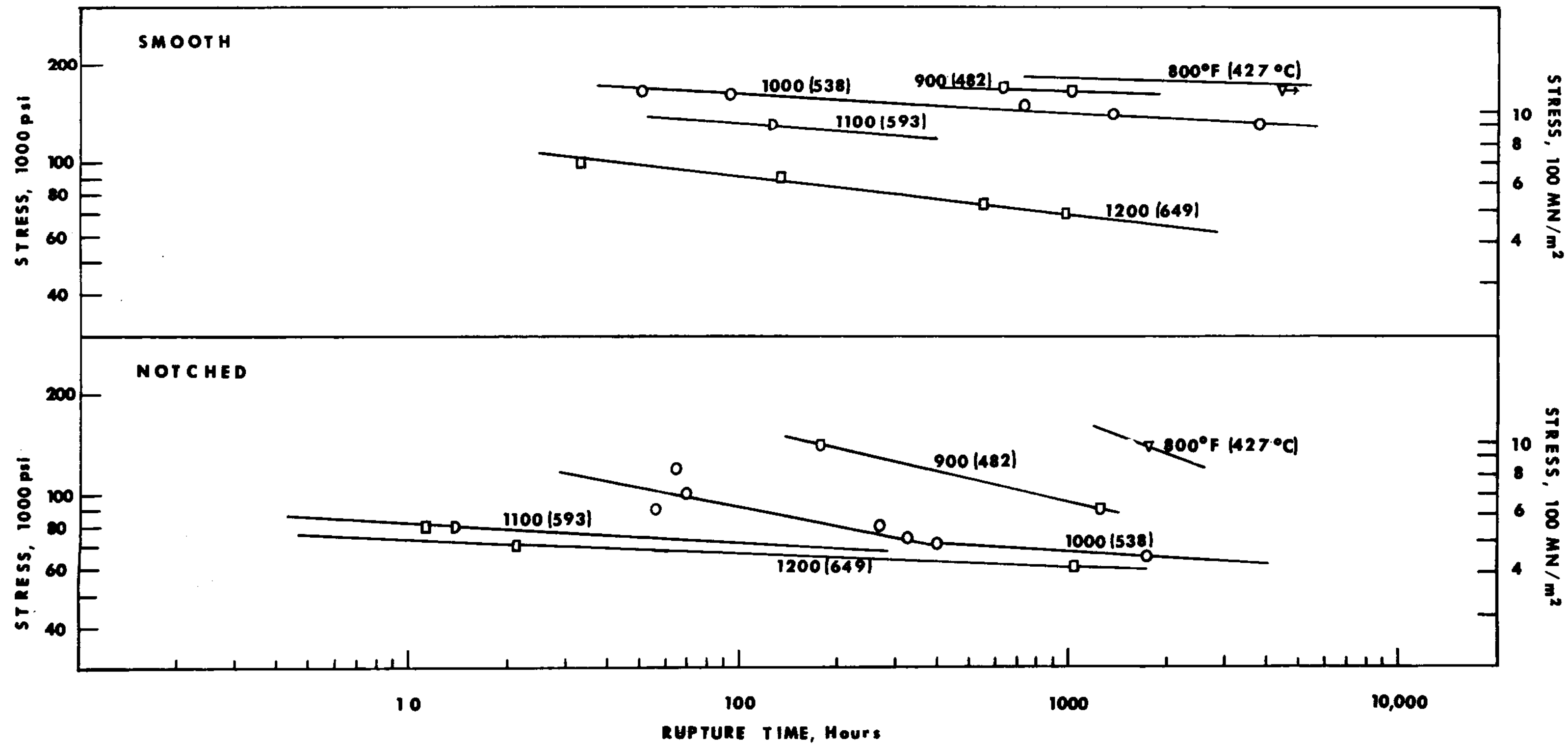


Figure 6. Stress versus rupture time data at temperatures from 800°F to 1200°F (427 - 649°C) obtained from smooth and notched specimens of Inconel 718 sheet heat treated 1 hour at 1750°F (954°C) and aged (ref. 2). Time-dependent notch sensitivity was evident at test temperatures from 900°F to 1100°F (482 -593°C). Little or no notch sensitivity was evident at 1200°F (649°C).

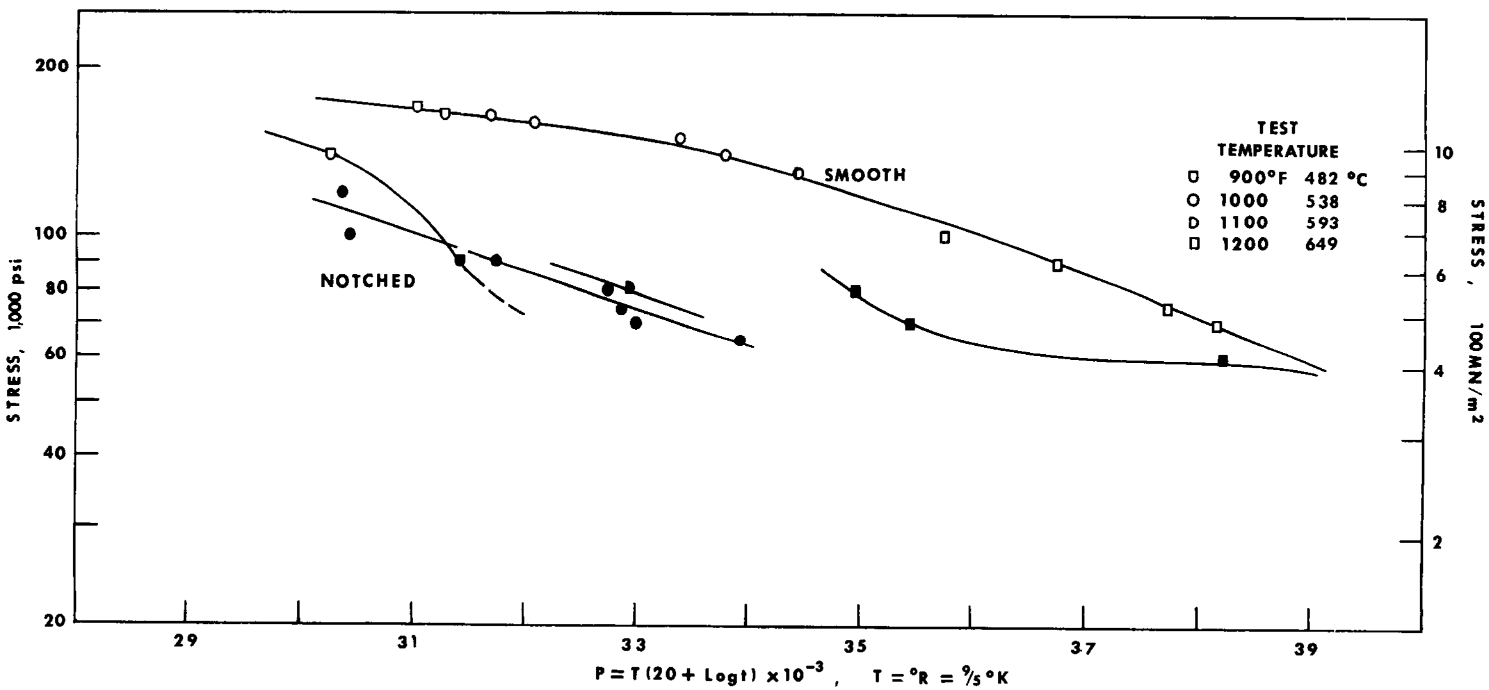


Figure 7. Time-temperature dependence of the rupture strengths of smooth and notched specimens of Inconel 718 sheet heat treated 1 hour at 1750°F (954°C) and aged.

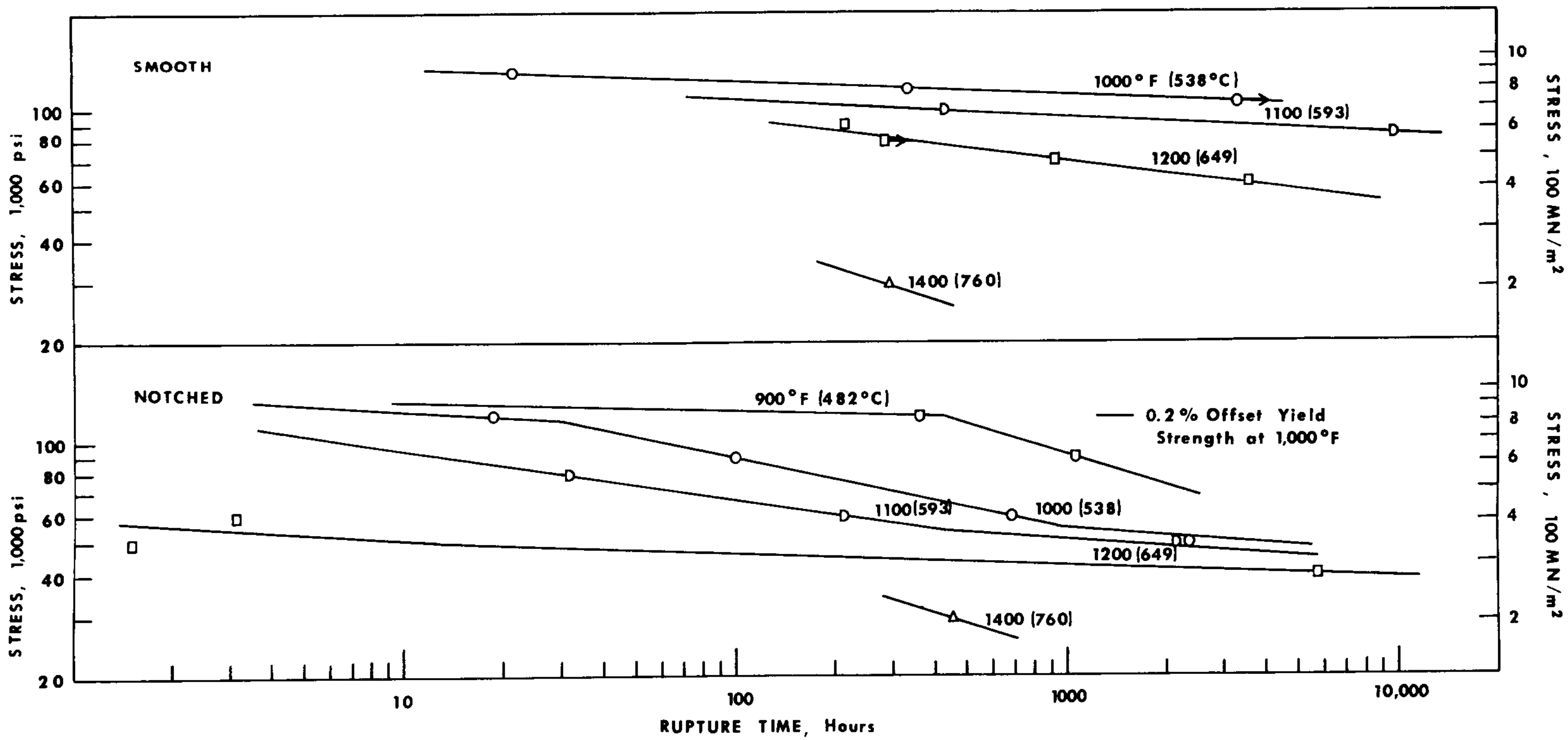


Figure 8. Stress versus rupture time data at temperatures from 900°F to 1400°F (482 - 760°C) obtained from smooth and notched specimens of Inconel 718 sheet heat treated 10 hours at 1950°F (1066°C) plus 48 hours at 1350°F (732°C). Time-dependent notch sensitivity was evident at temperatures from 900°F to 1200°F (482 - 649°C) but not at 1400°F (760°C).

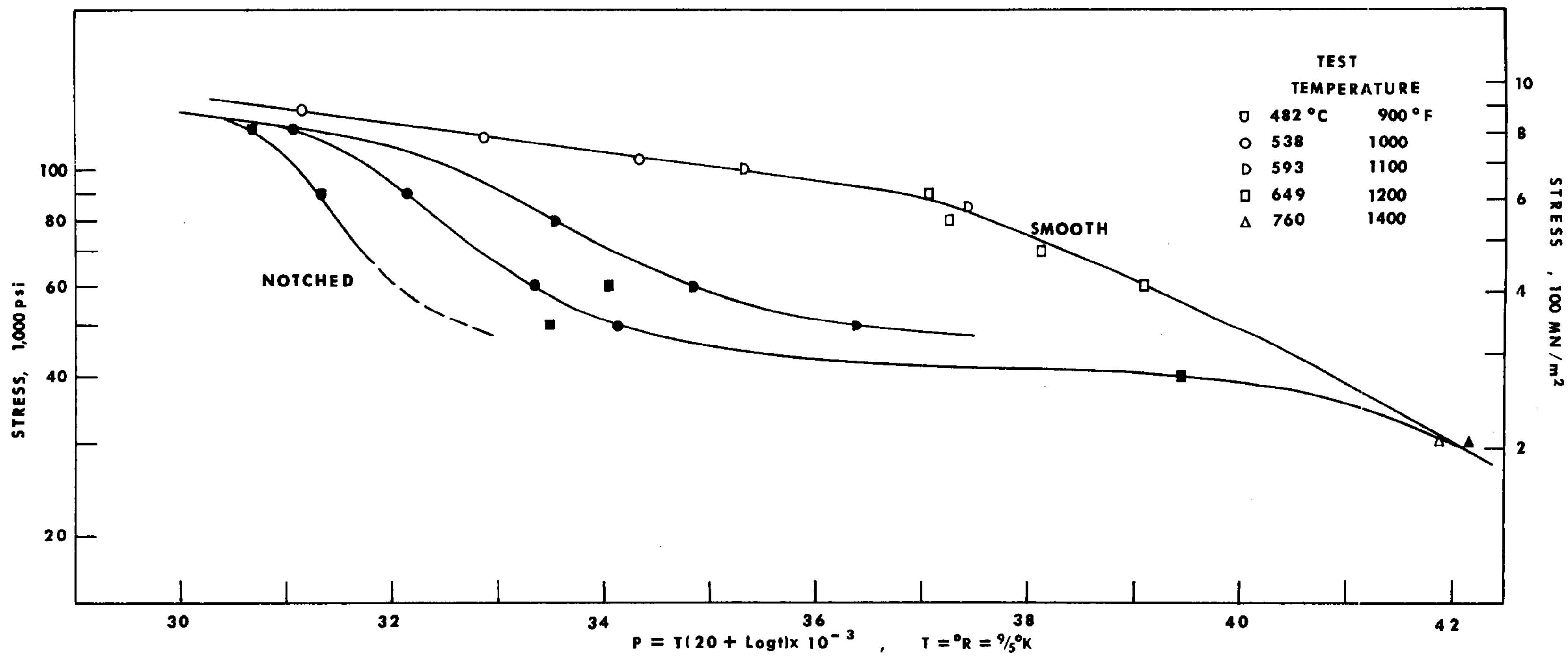


Figure 9. Time-temperature dependence of the rupture strengths of smooth and notched specimens of Inconel 718 sheet heat treated 10 hours at 1950°F (1066°C) plus 48 hours at 1350°F (732°C).

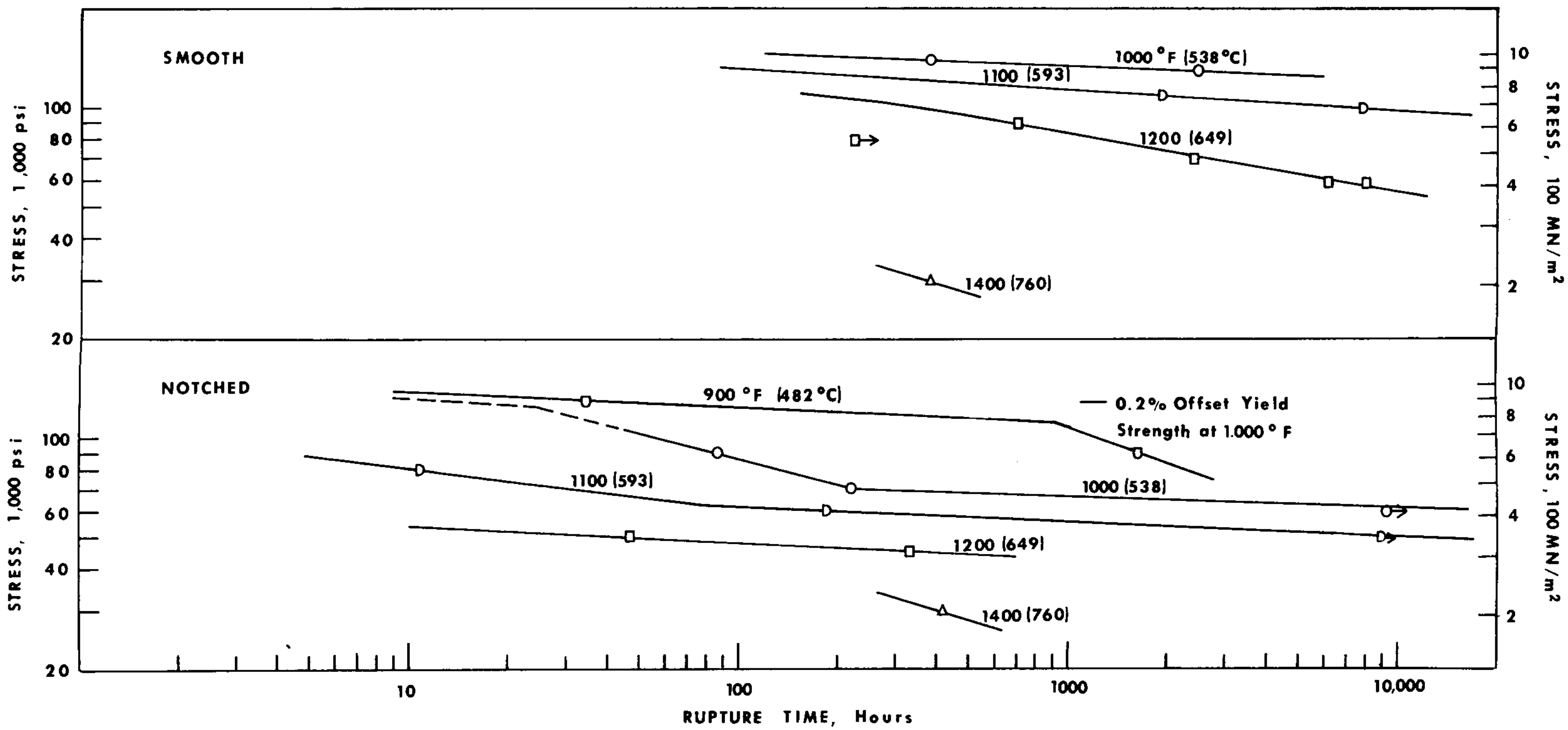


Figure 10. Stress versus rupture time data at temperatures from 900°F to 1400°F (482 - 760°C) obtained from smooth and notched specimens of Inconel 718 sheet heat treated 1 hour at 1950°F (1066°C) plus 48 hours at 1350°F (732°C). Time-dependent notch sensitivity occurred at temperatures from 900°F to 1200°F (482 - 649°C) but not at 1400°F (760°C).

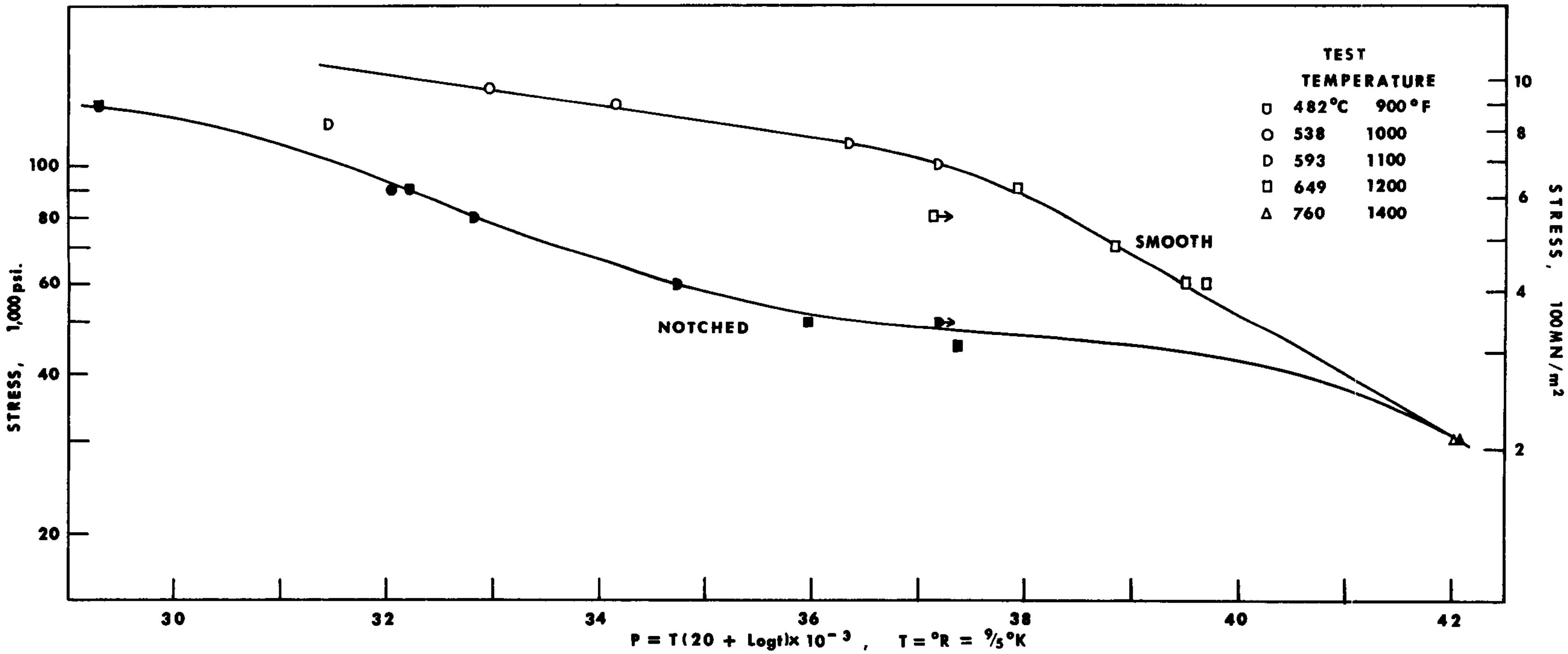


Figure 11. Time-temperature dependence of the rupture strengths of smooth and notched specimens of Inconel 718 sheet heat treated 1 hour at 1950°F (1066°C) plus 48 hours at 1350°F (732°C).

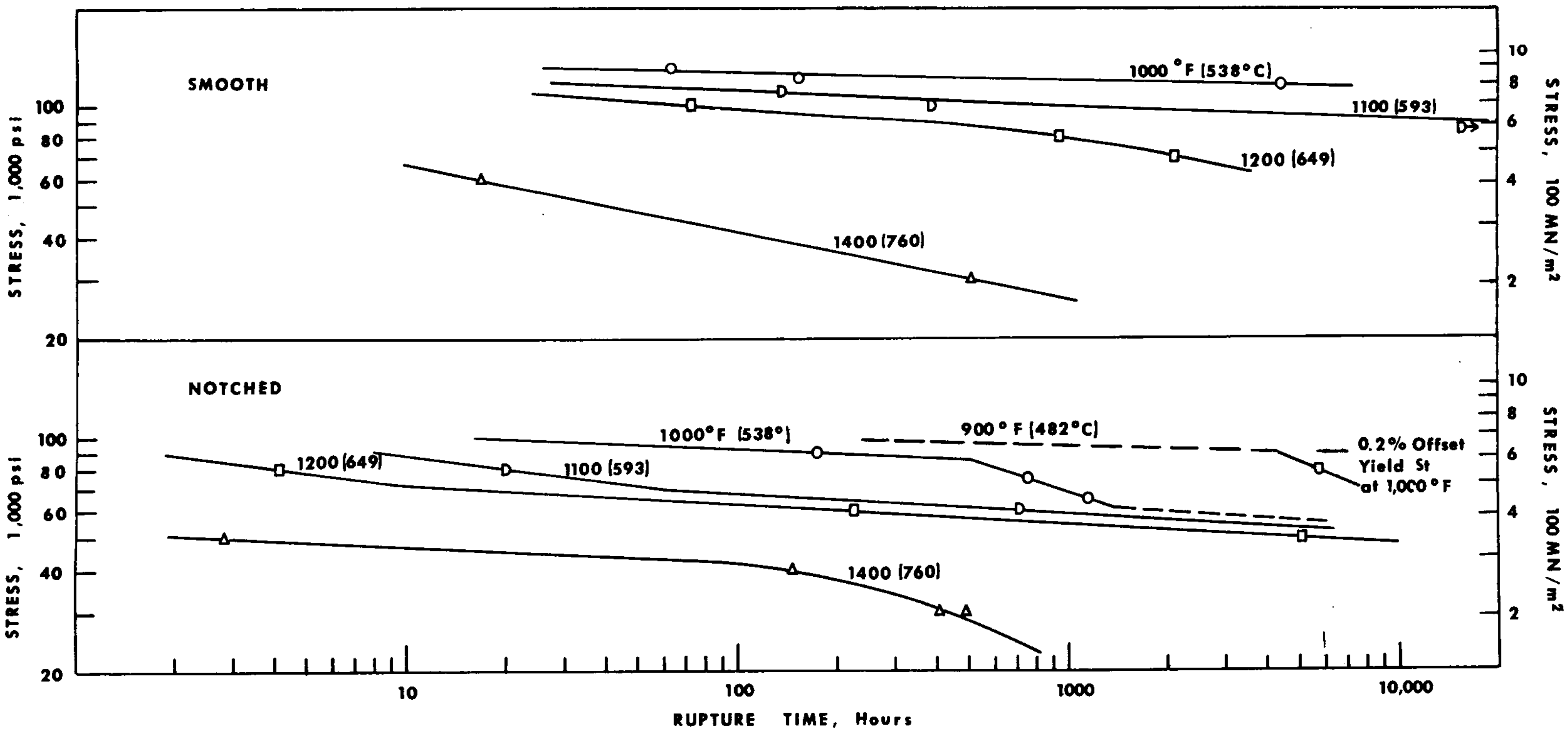


Figure 12. Stress versus rupture time data at temperatures from 900°F to 1400°F (482 - 760°C) obtained from smooth and notched specimens of Inconel 718 sheet heat treated 1 hour at 1950°F (1066°C) plus 2 hours at 1550°F (843°C). Time-dependent notch sensitivity was evident at temperatures from 900°F to 1200°F (482 - 649°C).

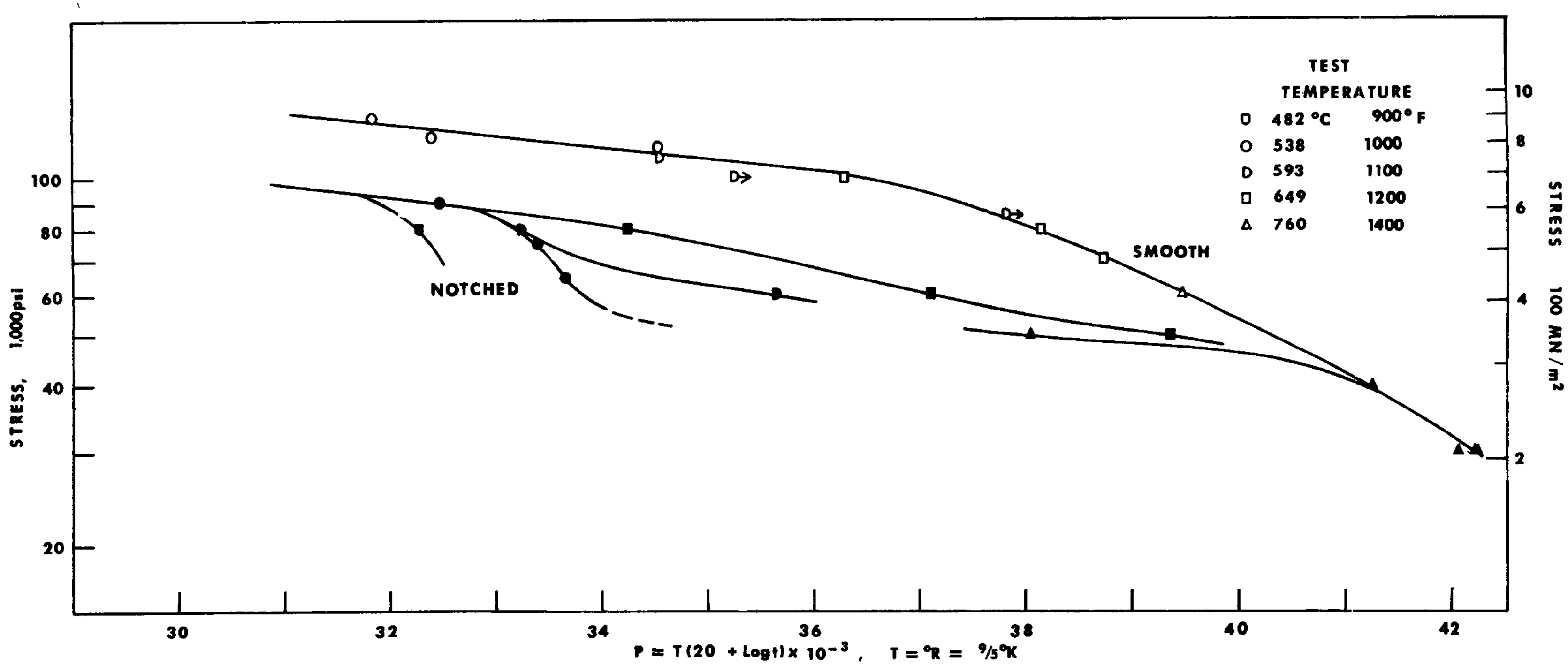


Figure 13. Time-temperature dependence of the rupture strengths of smooth and notched specimens of Inconel 718 sheet heat treated 1 hour at 1950°F (1066°C) plus 2 hours at 1550°F (843°C).

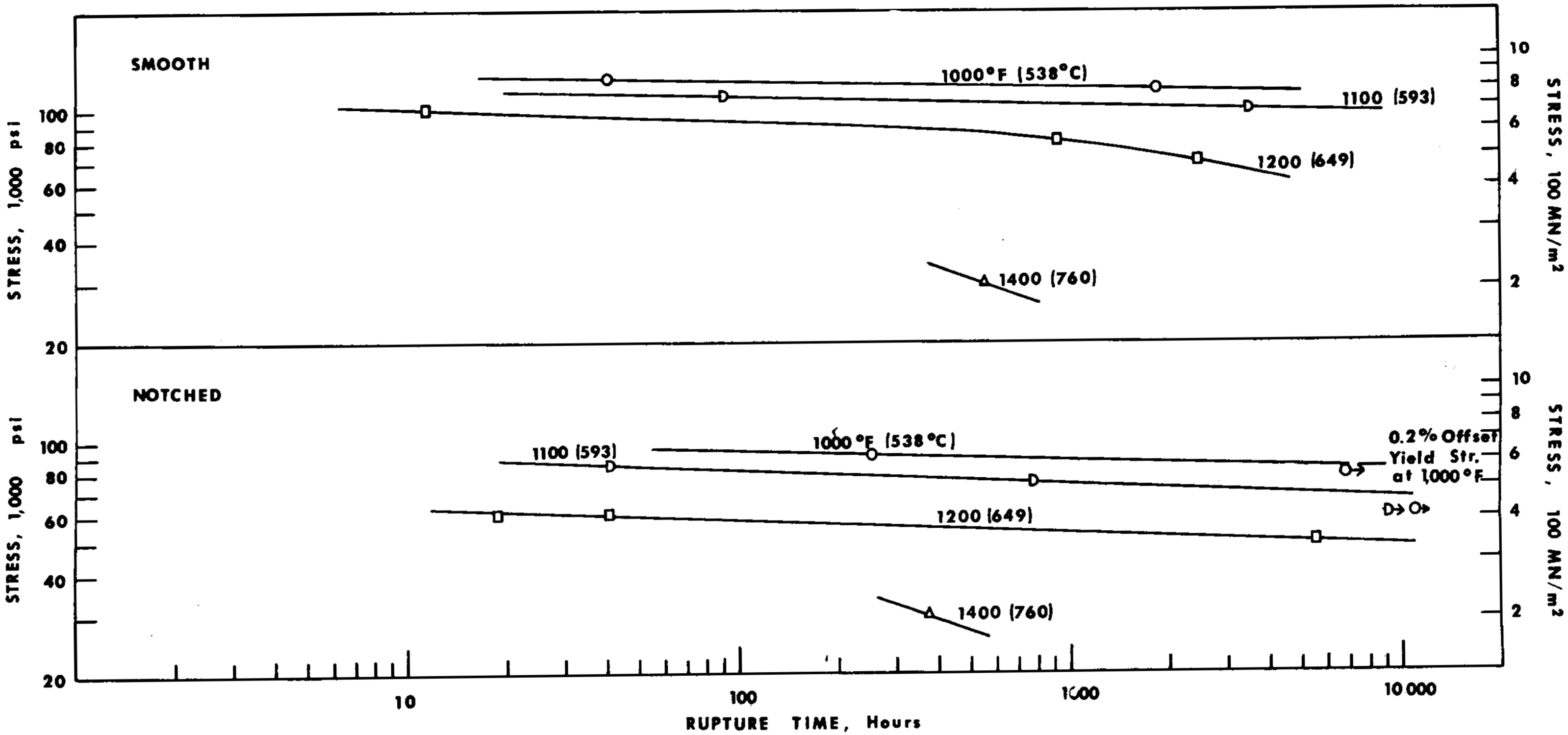


Figure 14. Stress versus rupture time data at temperatures from 1000°F to 1400°F (538 - 760°C) obtained from smooth and notched specimens of Inconel 718 sheet heat treated 1 hour at 1950°F (1066°C) plus 24 hours at 1550°F (843°C). The tests showed no time-dependent notch sensitivity.

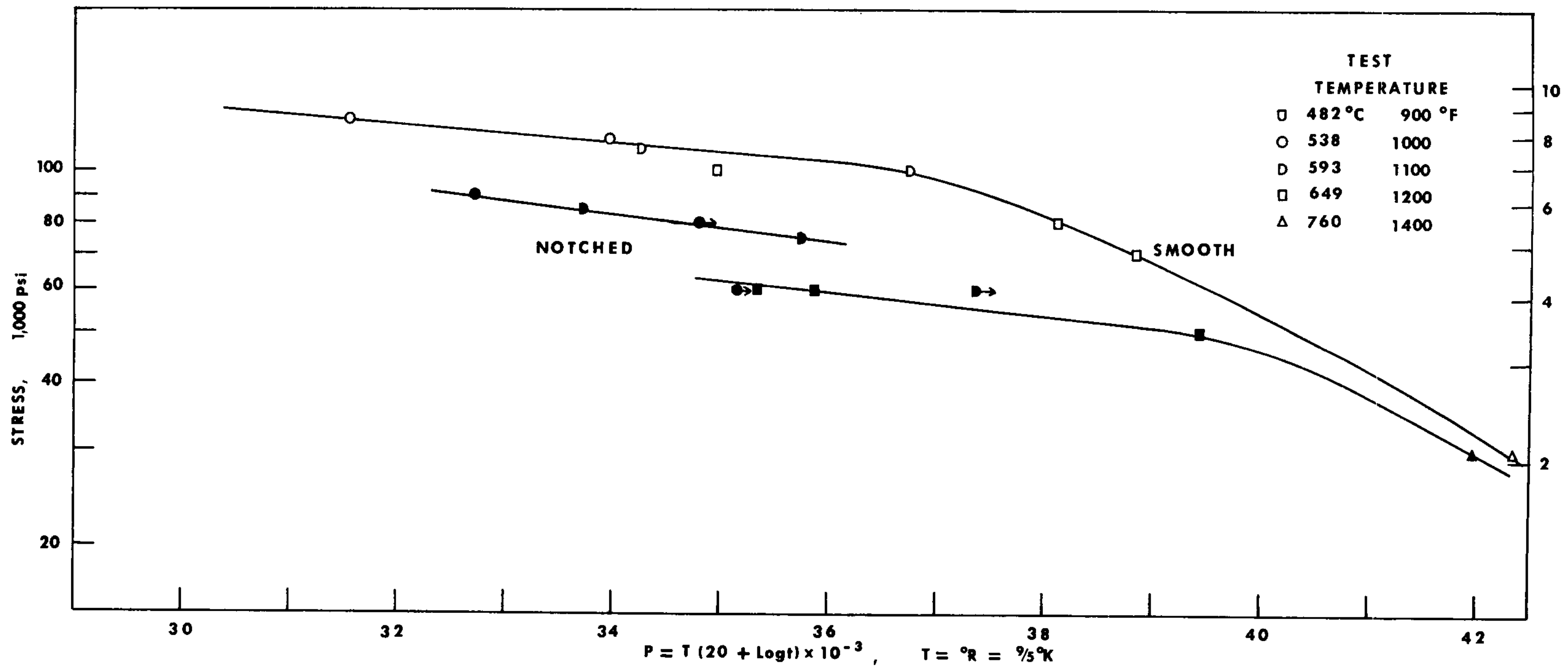


Figure 15. Time-temperature dependence of the rupture strengths of smooth and notched specimens of Inconel 718 sheet heat treated 1 hour at 1950°F (1066°C) plus 24 hours at 1550°F (843°C).

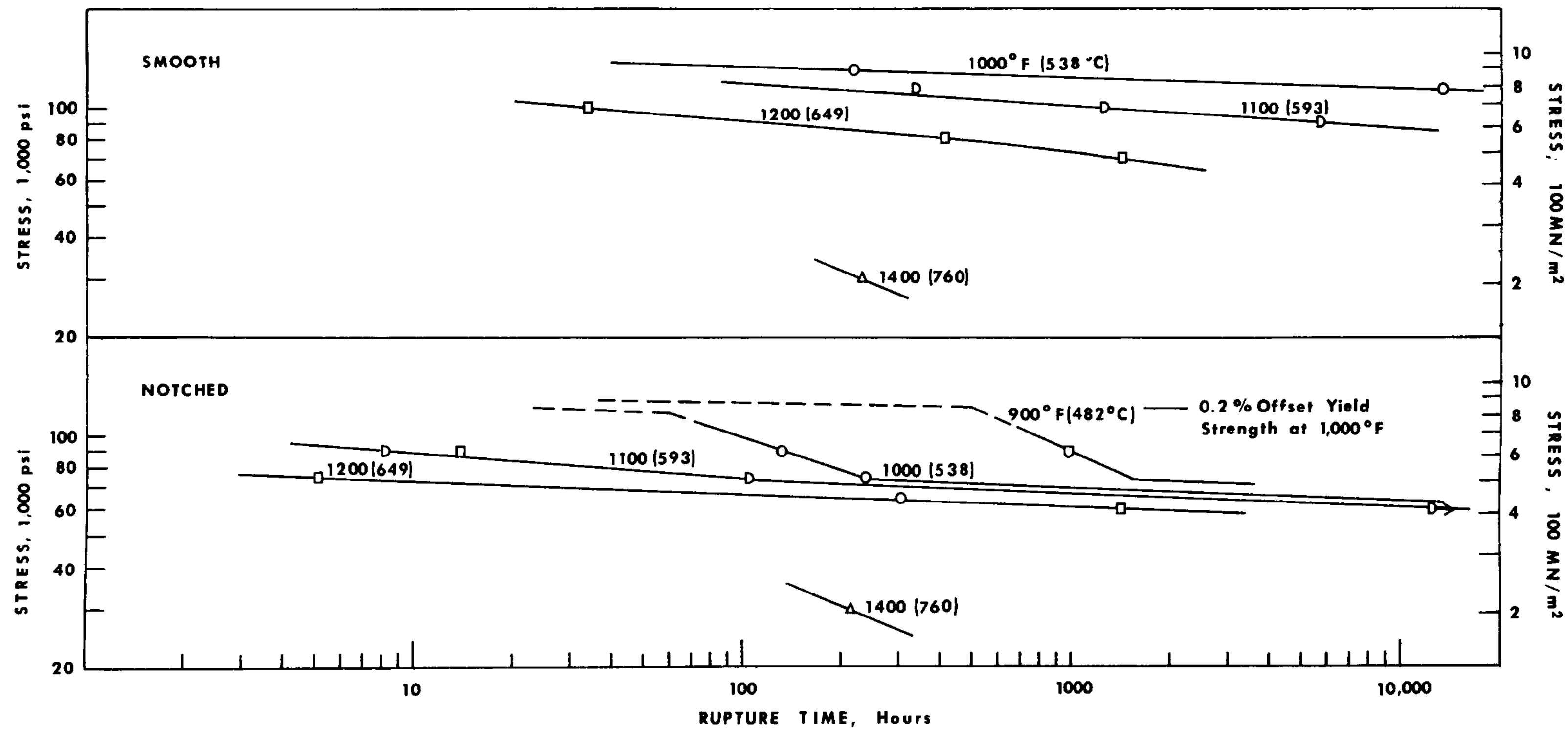


Figure 16. Stress versus rupture time data at temperatures from 900°F to 1400°F (482 - 760°C) obtained from smooth and notched specimens of Inconel 718 sheet heat treated 10 hours at 1800°F (982°C) plus 48 hours at 1350°F (732°C). Time-dependent notch sensitivity occurred at 900°F to 1100°F (482 - 593°C) but not at 1400°F (760°C).

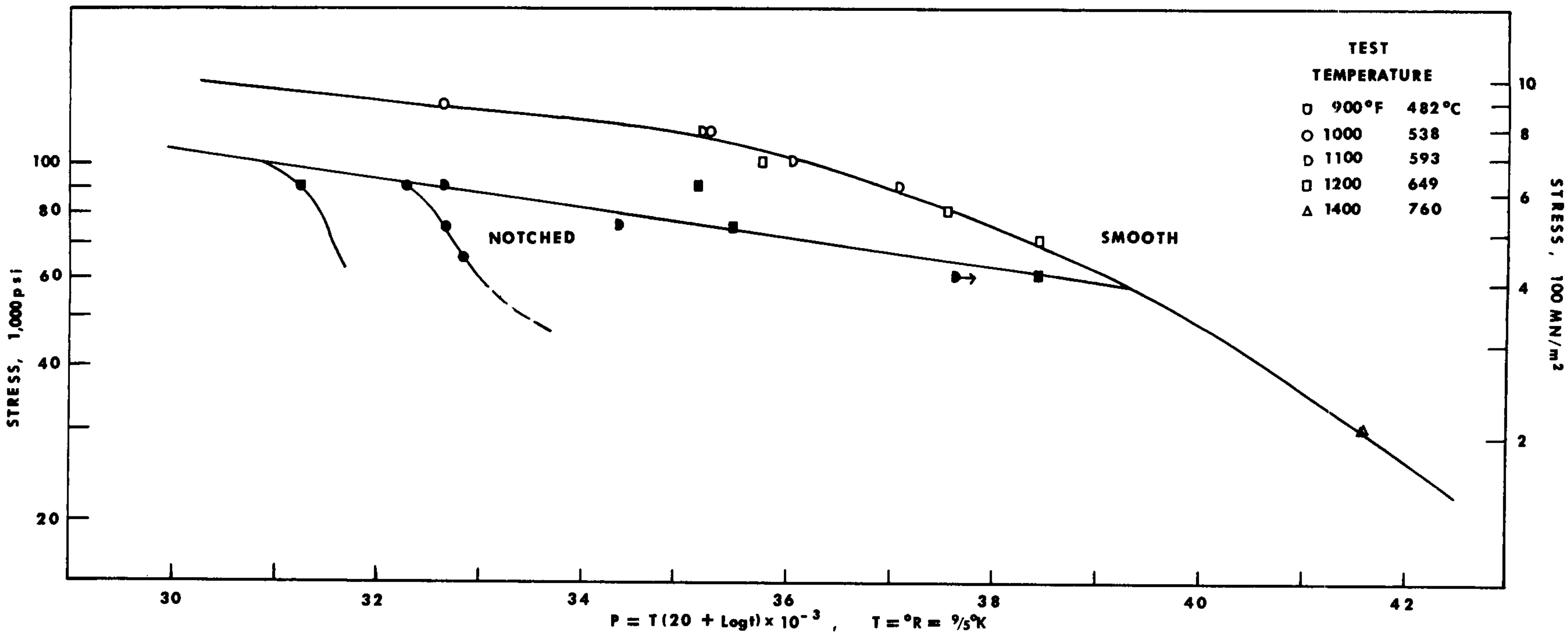


Figure 17. Time-temperature dependence of the rupture strengths of smooth and notched specimens of Inconel 718 sheet heat treated 10 hours at 1800°F(982°C) plus 48 hours at 1350°F (732°C).

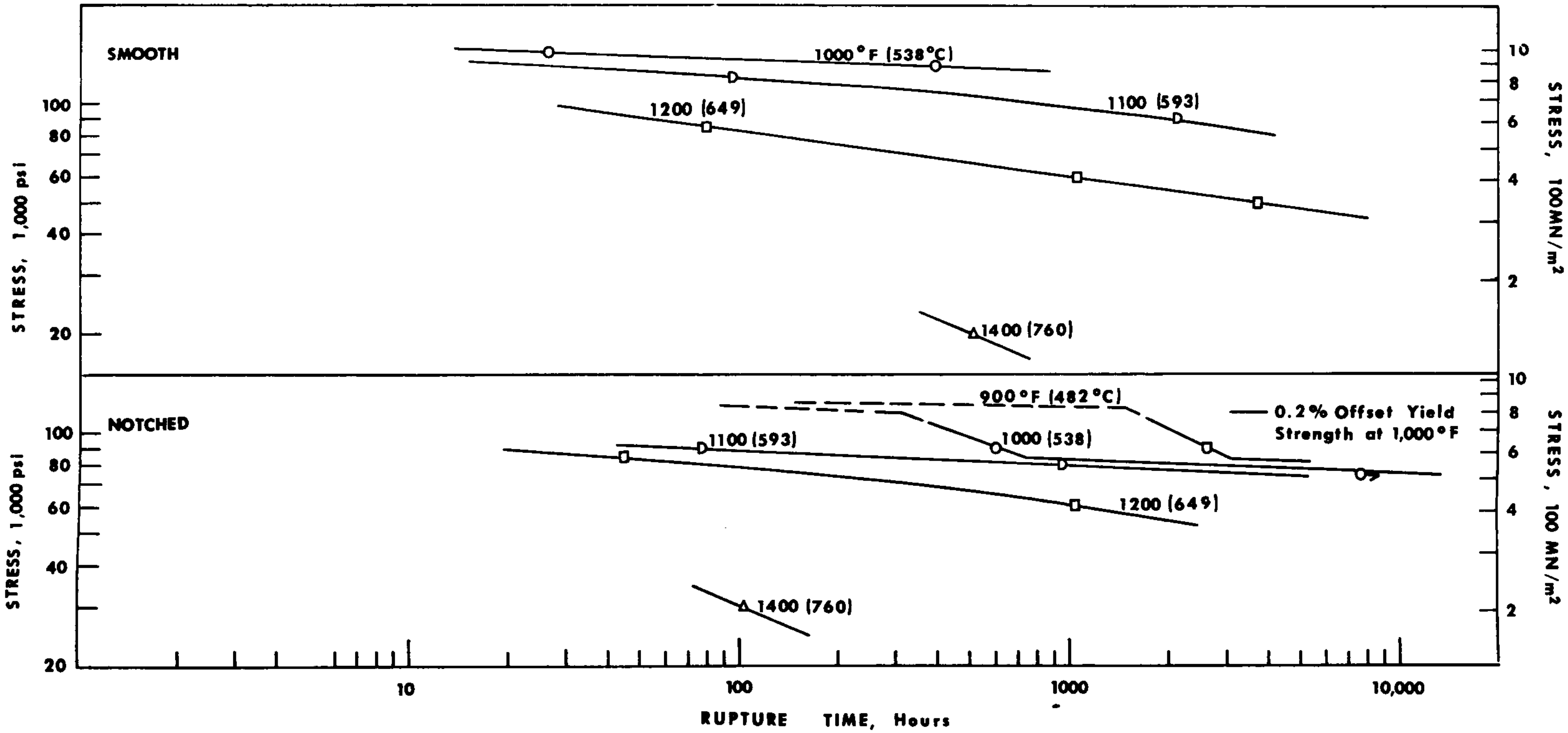


Figure 18. Stress versus rupture time data at temperatures from 900°F to 1400°F (482 - 760°C) obtained from smooth and notched specimens of Inconel 718 sheet heat treated 10 hours at 1700°F (927°C) plus 3 hours at 1325°F (718°C). Time-dependent notch sensitivity was observed at 900°F (482°C) and 1000°F (538°C), but not at 1200°F to 1400°F (649 - 760°C).

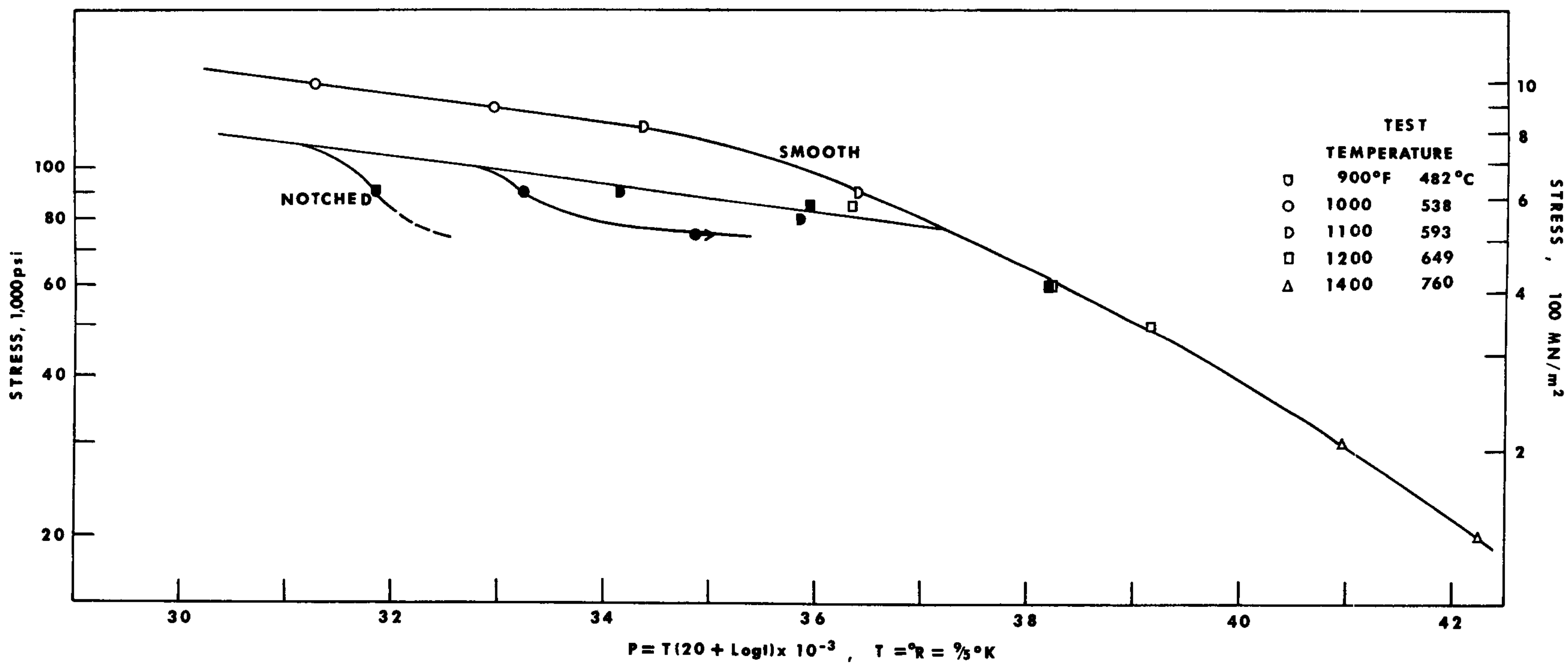


Figure 19. Time-temperature dependence of the rupture strengths of smooth and notched specimens of Inconel 718 sheet heat treated 10 hours at 1700°F (927°C) plus 3 hours at 1325°F (718°C).

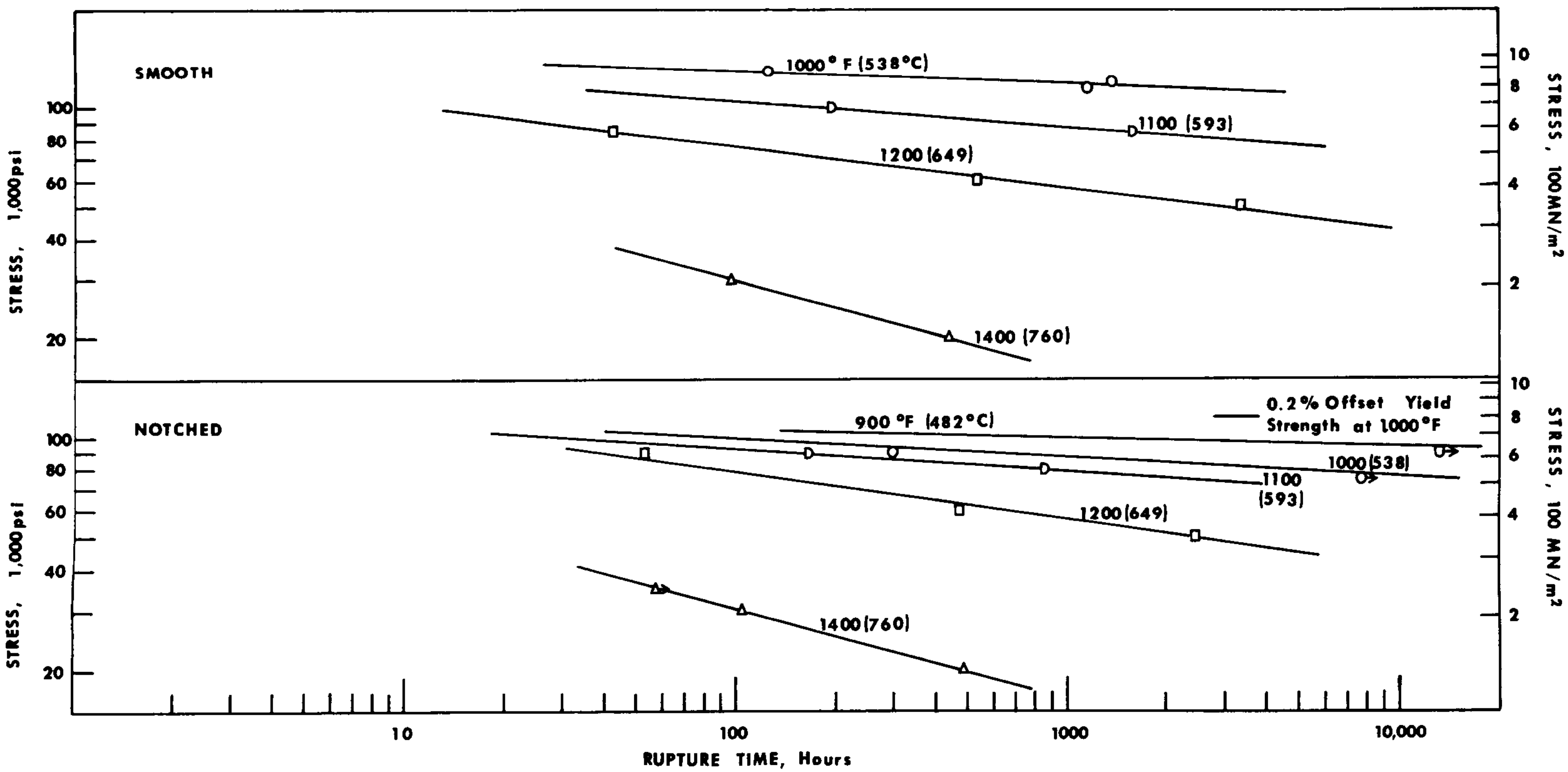


Figure 20. Stress versus rupture time data at temperatures from 900°F to 1400°F (482 - 760°C) obtained from smooth and notched specimens of Inconel 718 sheet heat treated 10 hours at 1700°F (927°C) plus 48 hours at 1350°F (732°C). The tests showed no time-dependent notch sensitivity.

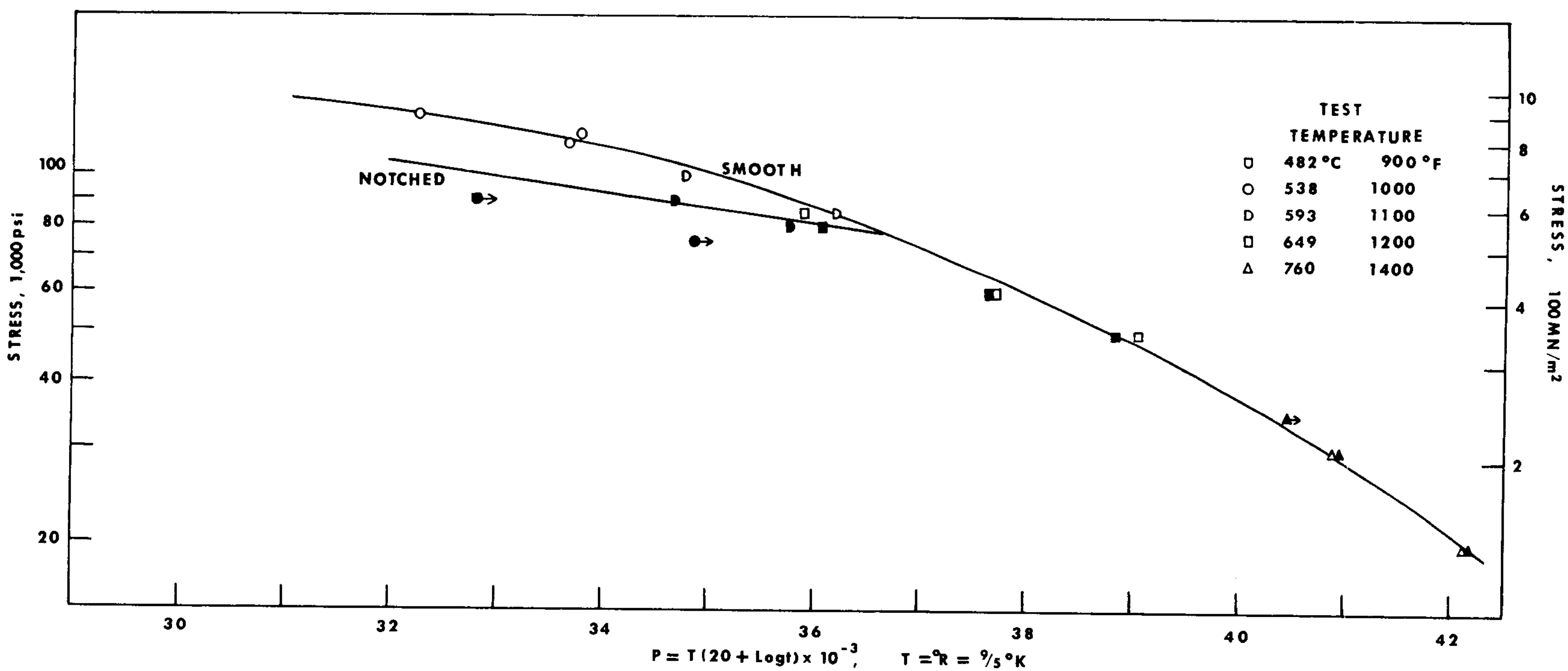


Figure 21. Time-temperature dependence of the rupture strengths of smooth and notched specimens of Inconel 718 sheet heat treated 10 hours at 1700°F (927°C) plus 48 hours at 1350°F (732°C)

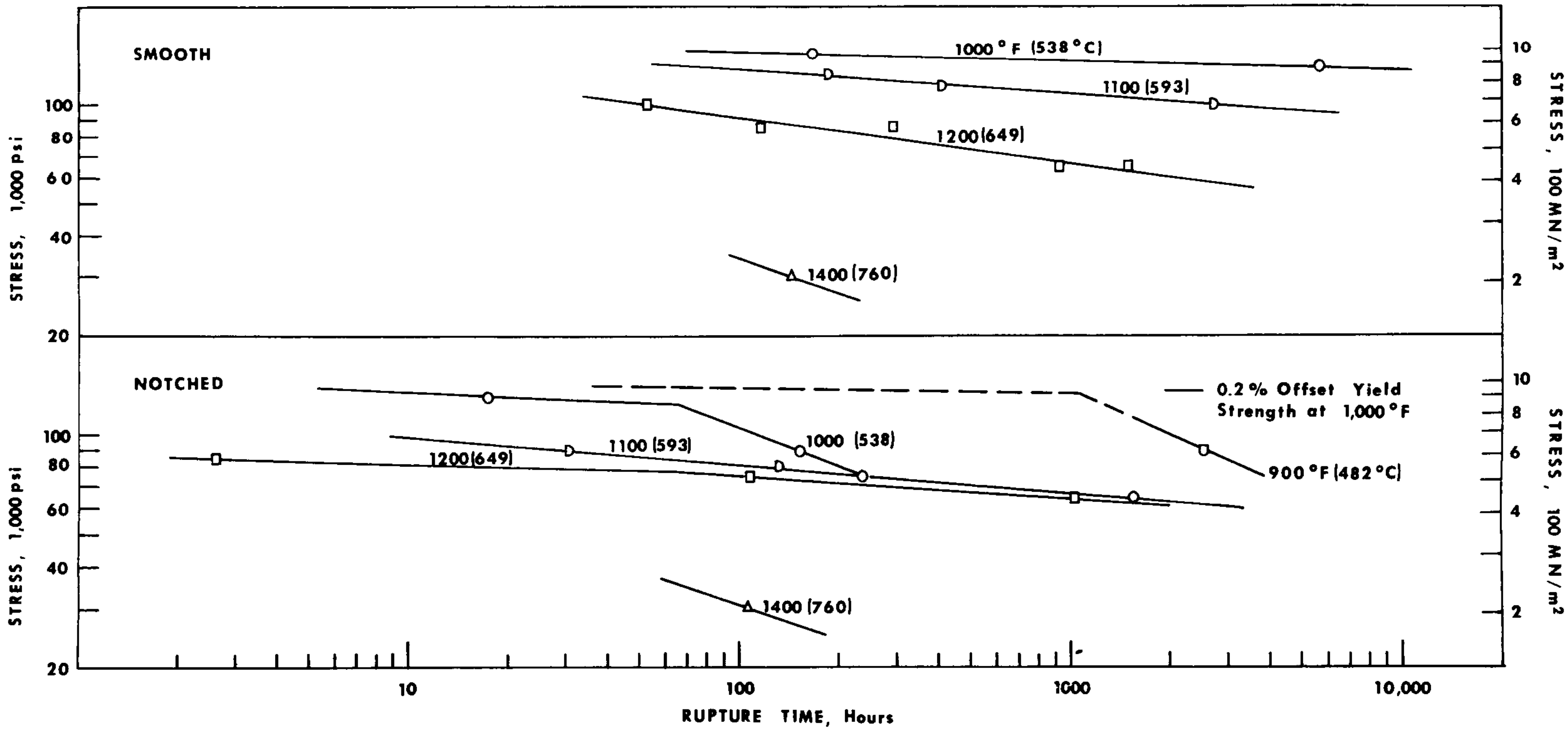


Figure 22. Stress versus rupture time data at temperatures from 900°F to 1400°F (482 - 760°C) obtained from smooth and notched specimens of Inconel 718 sheet heat treated 1 hour at 1700°F (927°C) plus 3 hours at 1325°F (718°C). Time-dependent notch sensitivity was observed at 900°F to 1100°F (482 - 593°C) but not at 1400°F (760°C).

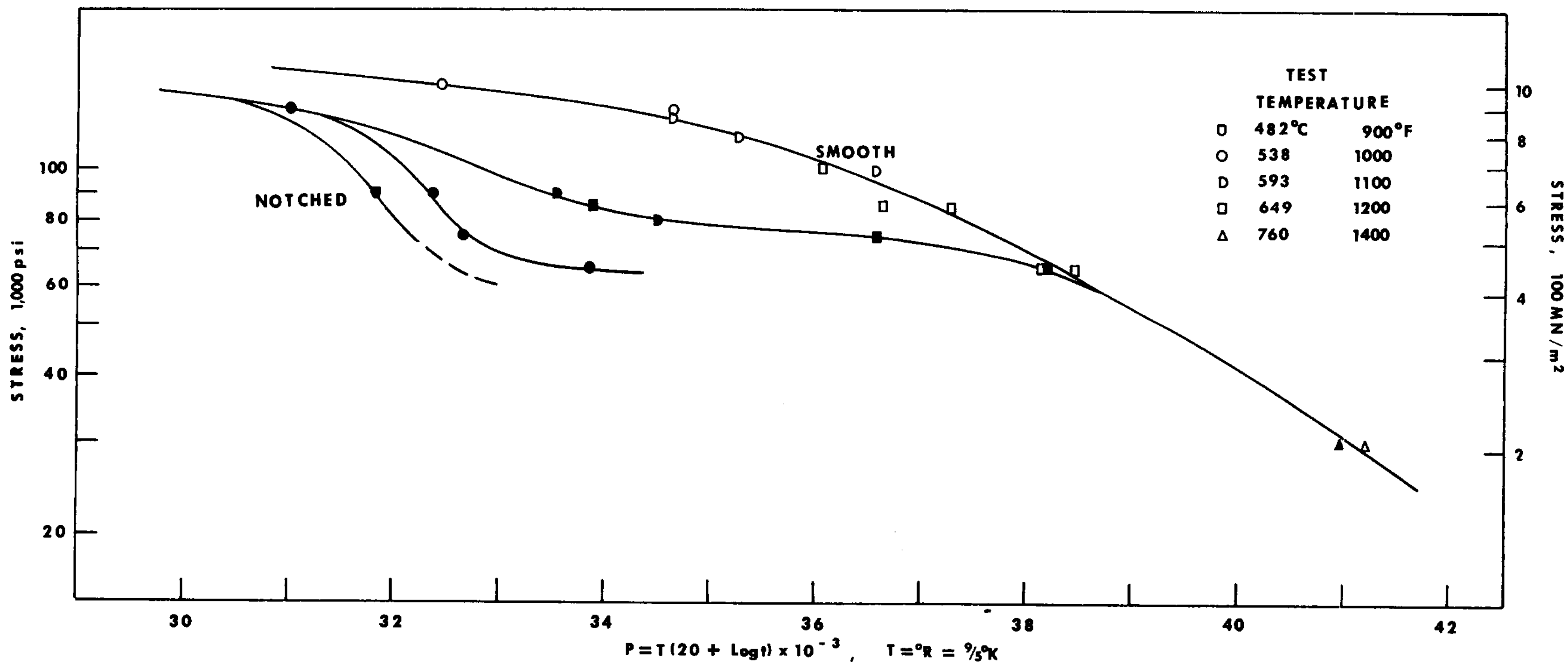


Figure 23. Time-temperature dependence of the rupture strengths of smooth and notched specimens of Inconel 718 sheet heat treated 1 hour at 1700°F (927°C) plus 3 hours at 1325°F (718°C).

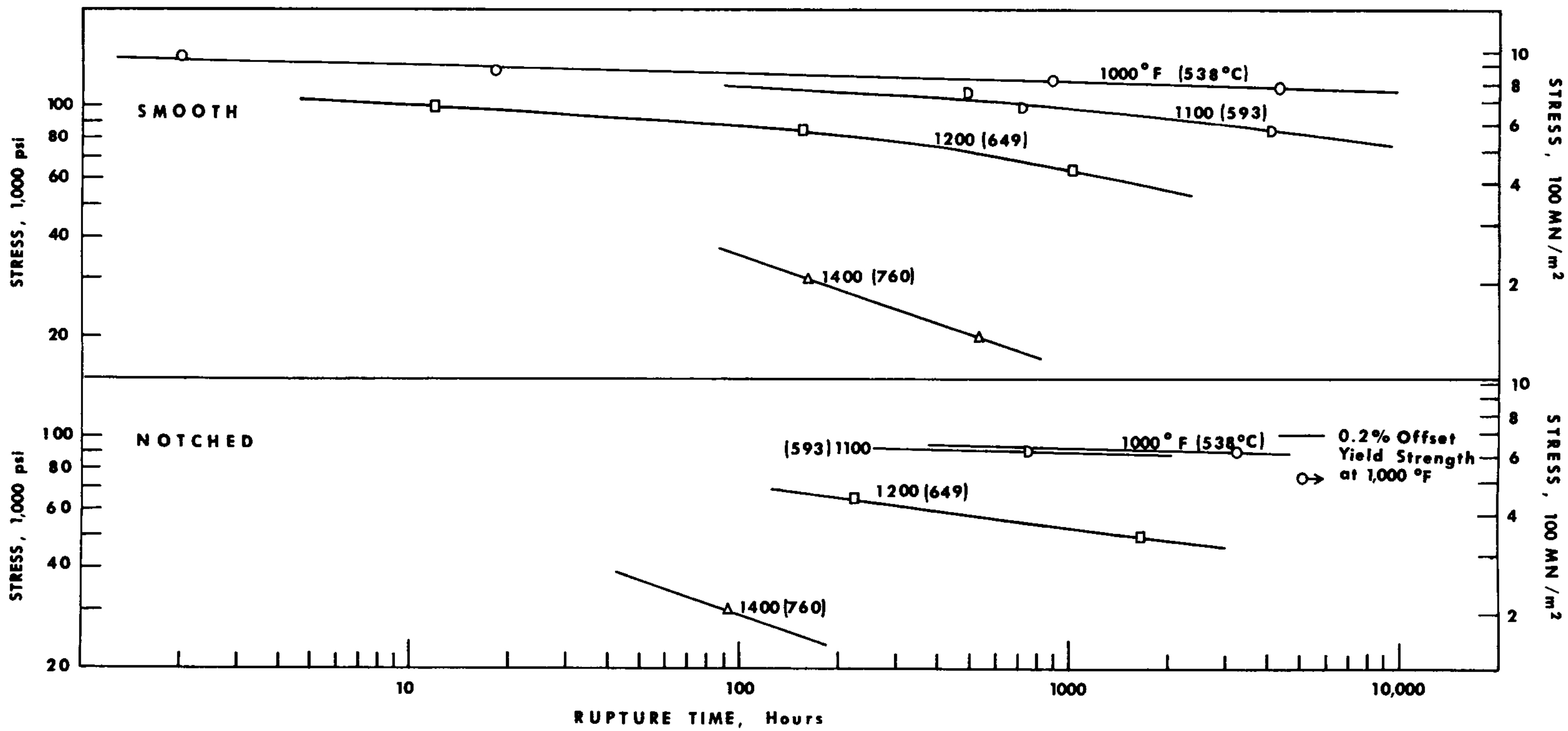


Figure 24. Stress versus rupture time data at temperatures from 1000°F to 1400°F (538 - 760°C) obtained from smooth and notched specimens of Inconel 718 sheet heat treated 1 hour at 1700°F (927°C) plus 2 hours at 1550°F (843°C). Time-dependent notch sensitivity did not occur.

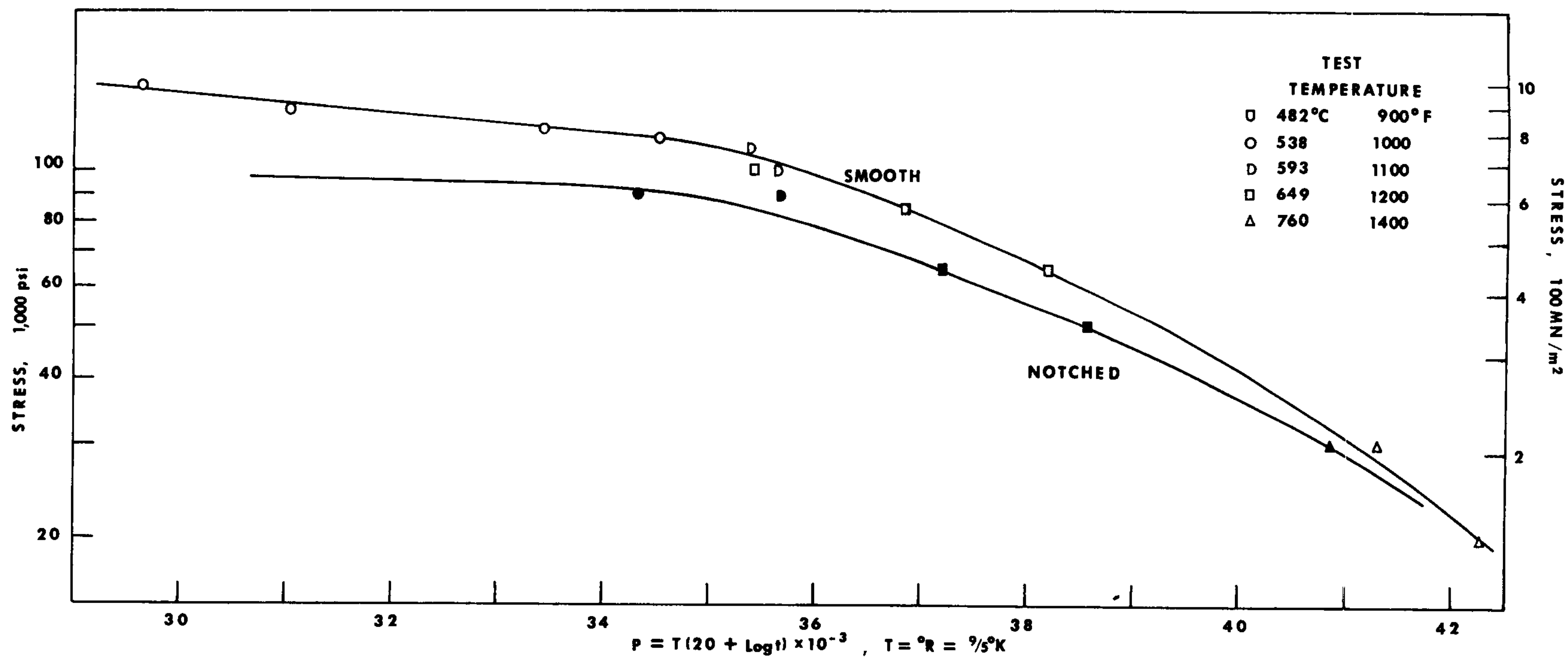


Figure 25. Time-temperature dependence of the rupture strengths of smooth and notched specimens of Inconel 718 sheet heat treated 1 hour at 1700°F (927°C) plus 2 hours at 1550°F (843°C).

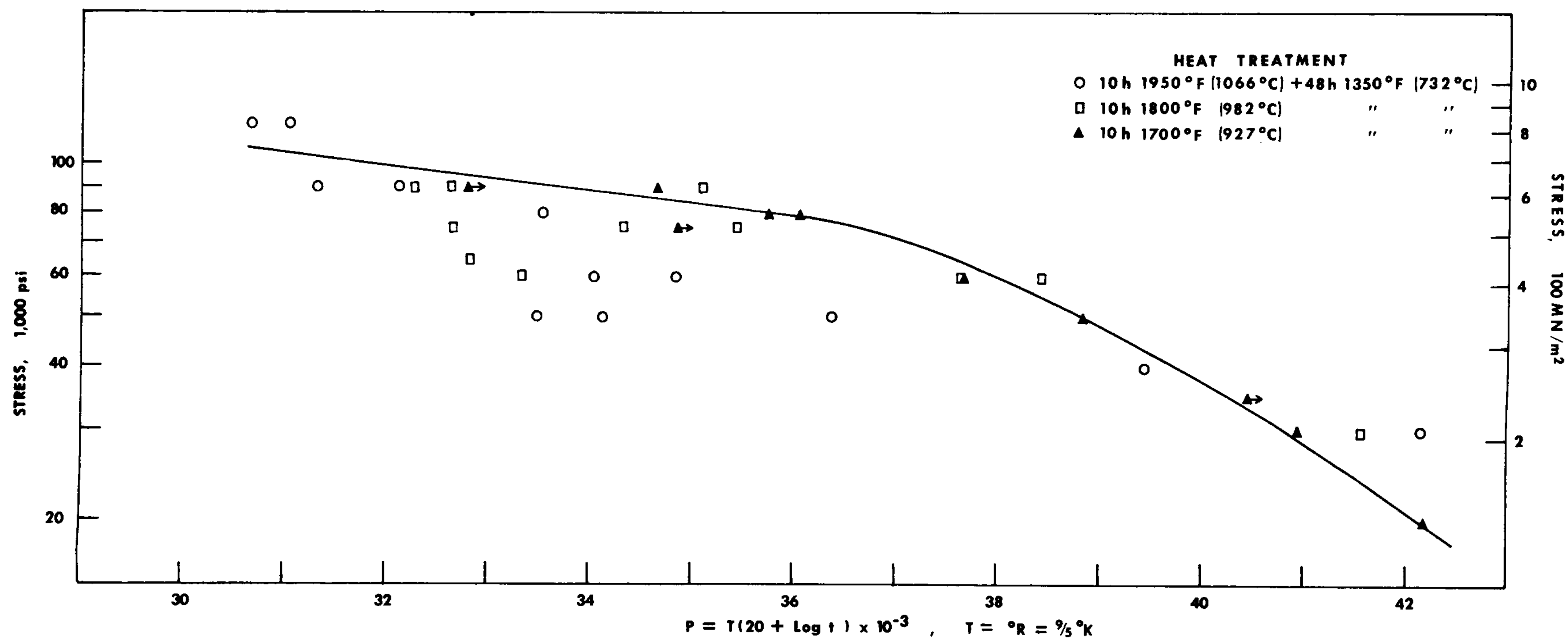


Figure 26. Time-temperature dependence of the rupture strengths at 900° to 1400°F (482 - 760°C) for notched specimens of Inconel 718 in various heat treated conditions.

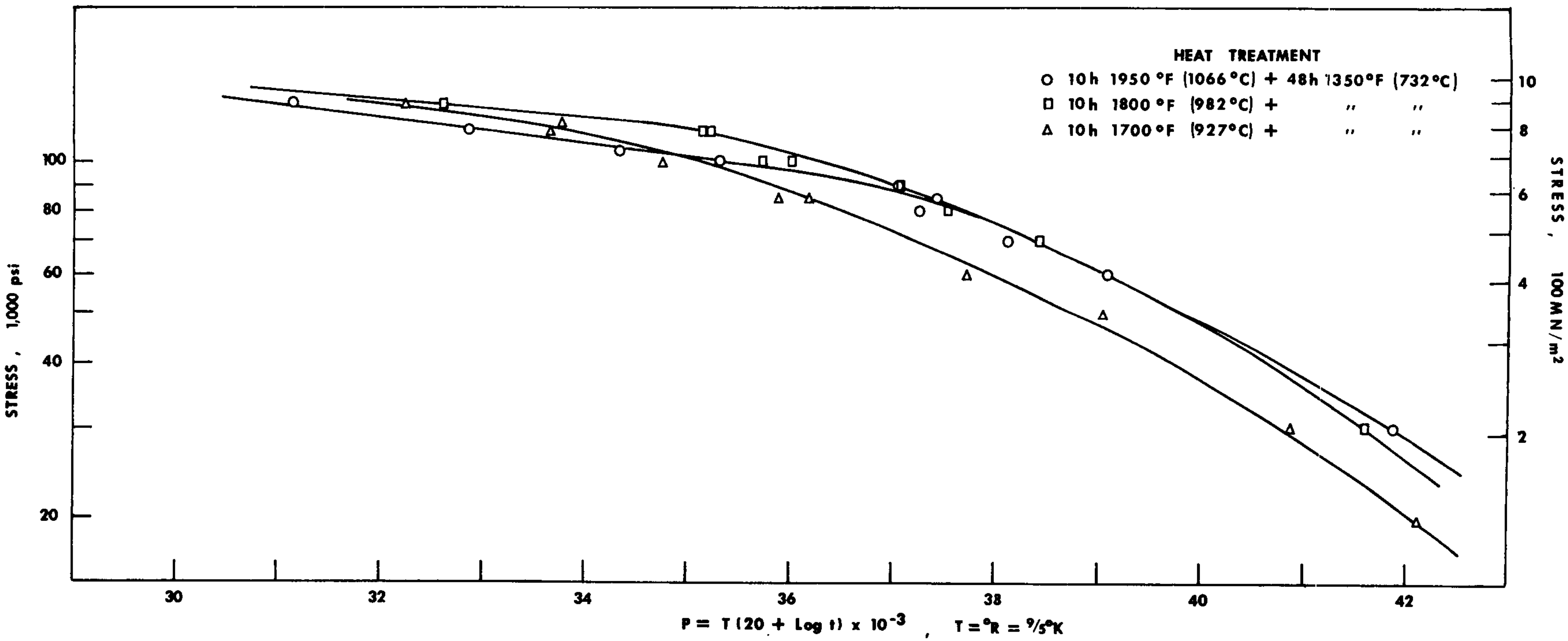


Figure 27. Time-temperature dependence of the rupture strengths at 900° to 1400°F (482 - 760°C) for smooth specimens of Inconel 718 in various heat treated conditions.

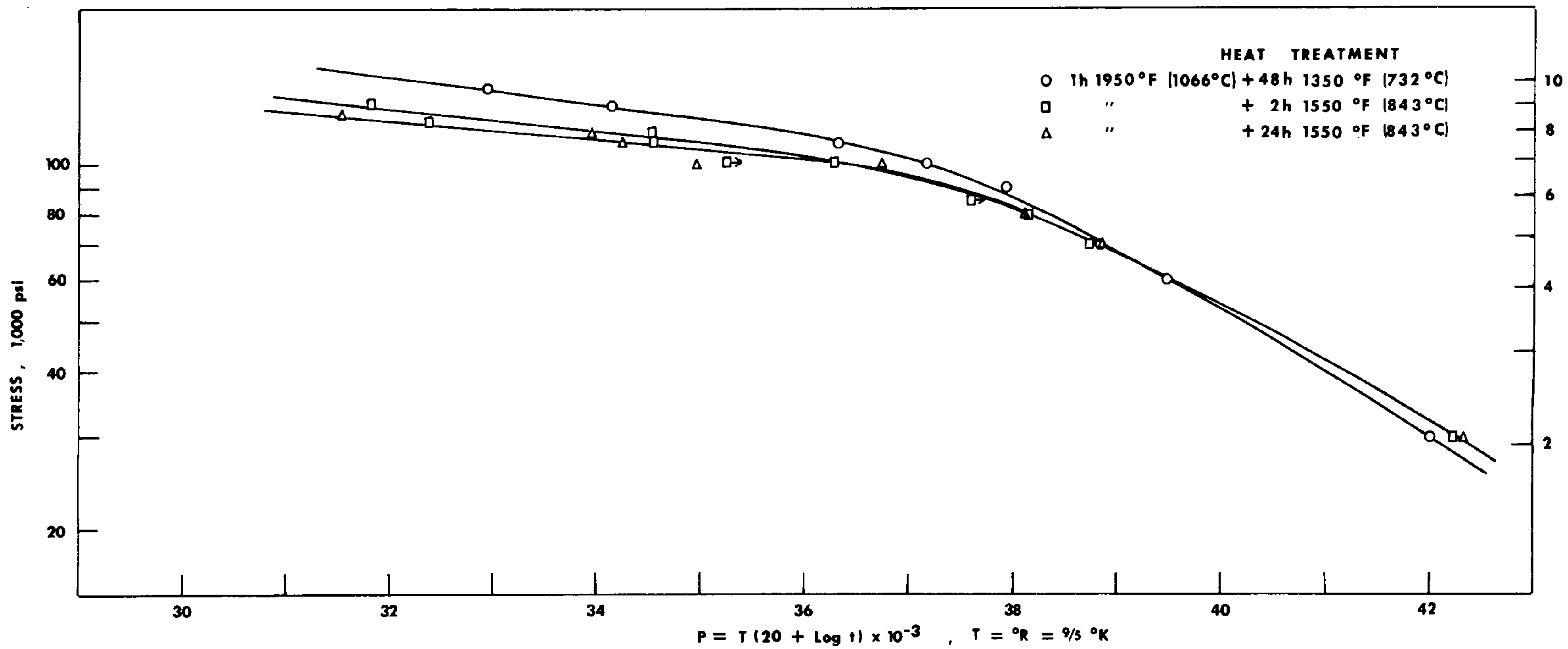


Figure 28. Time-temperature dependence of the rupture strengths at 900° to 1400°F (482 - 760°C) for smooth specimens of Inconel 718 in various heat treated conditions.

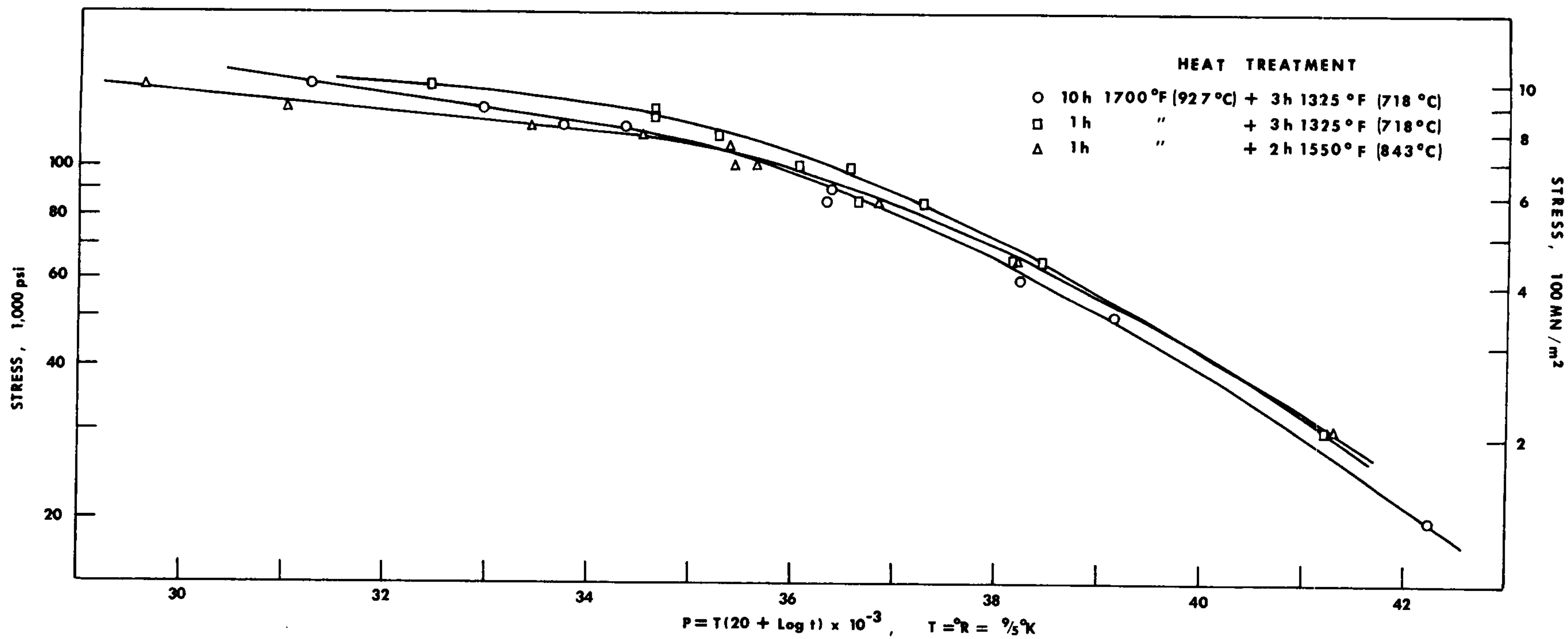
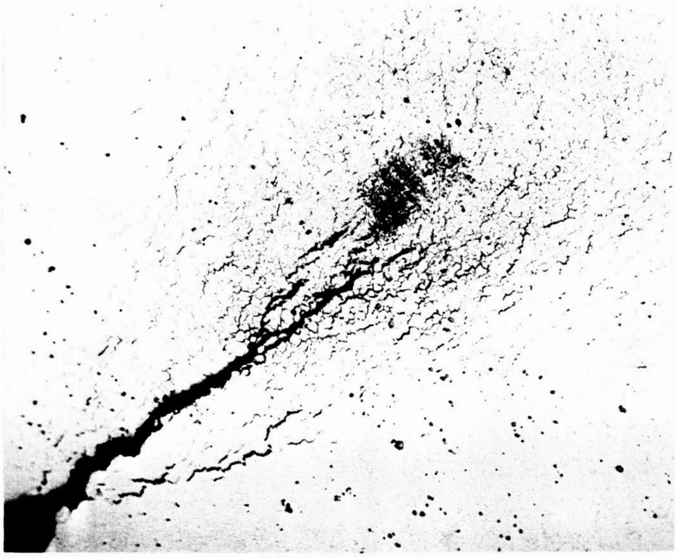


Figure 29. Time-temperature dependence of the rupture strengths at 900° to 1400°F (482 - 760°C) for smooth specimens of Inconel 718 in various heat treated conditions.

(a) 75x



(b) 75x

Figure 30. Optical photomicrographs of notched specimens of Inconel 718 polished to remove the oxidized surface layers. Microcracks in (a) are an early stage of crack initiation (heat treated 10 hours at 1800°F (982°C) plus 48 hours at 1350°F (732°C), tested at 1100°F (593°C) at 60ksi (414MN/m^2), discontinued after 12,867 hours). The "through the thickness" crack in (b) is a late stage of crack growth (heat treated 10 hours at 1700°F (927°C) plus 48 hours at 1350°F (732°C), tested at 1000°F(538°C) at 75ksi(517MN/m^2), discontinued after 7656 hours).

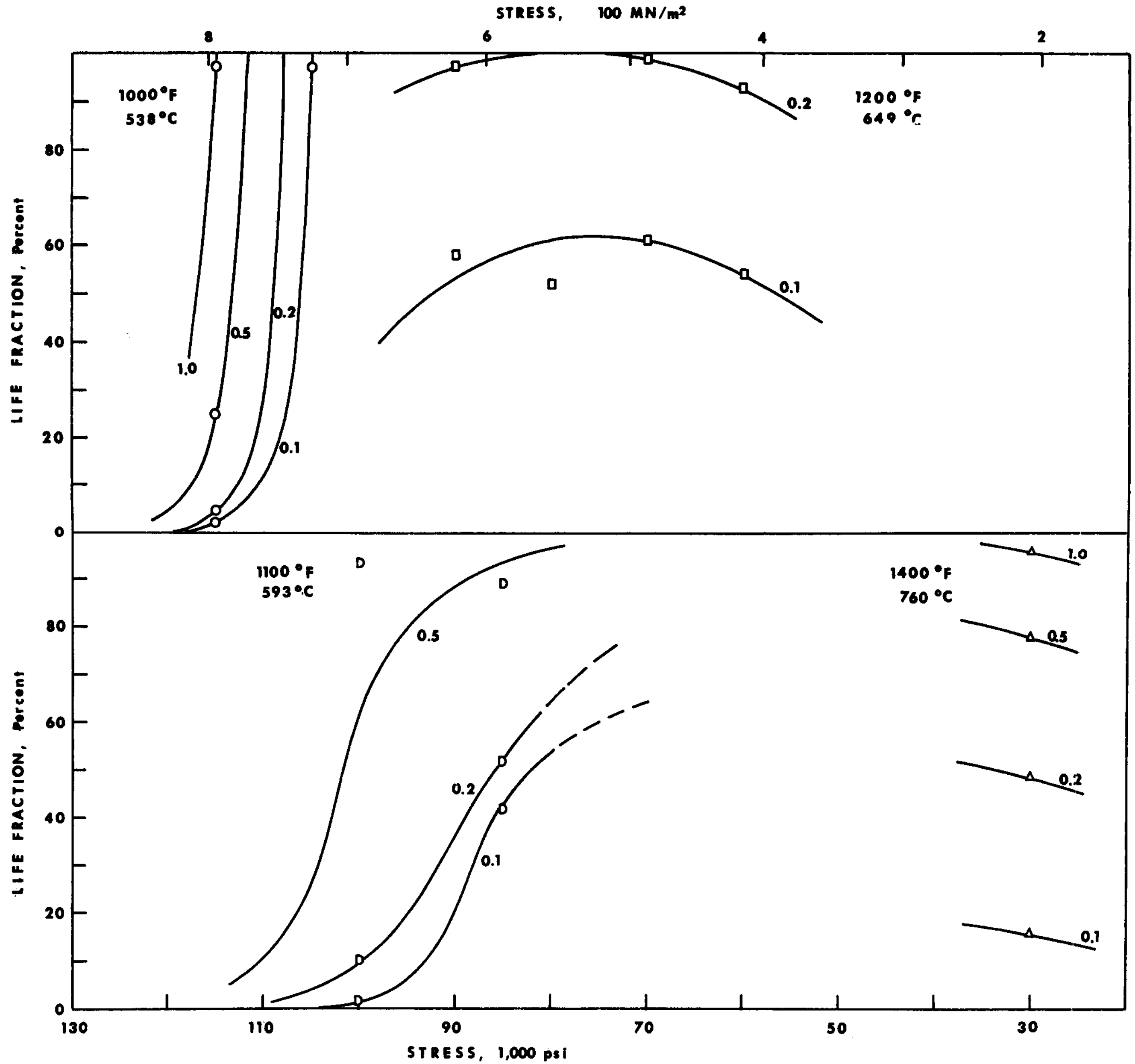


Figure 31. Iso-creep strain curves of life fraction versus stress at temperatures from 1000°F (538°C) to 1400°F (760°C) for Inconel 718 heat treated 10 hours at 1950°F (1066°C) plus 48 hours at 1350°F (732°C). Time-dependent notch sensitivity occurred under test conditions where large amounts of rupture life were utilized for small creep strains at test temperatures 1000° 1100° and 1200°F (538°, 593°, and 649°C).

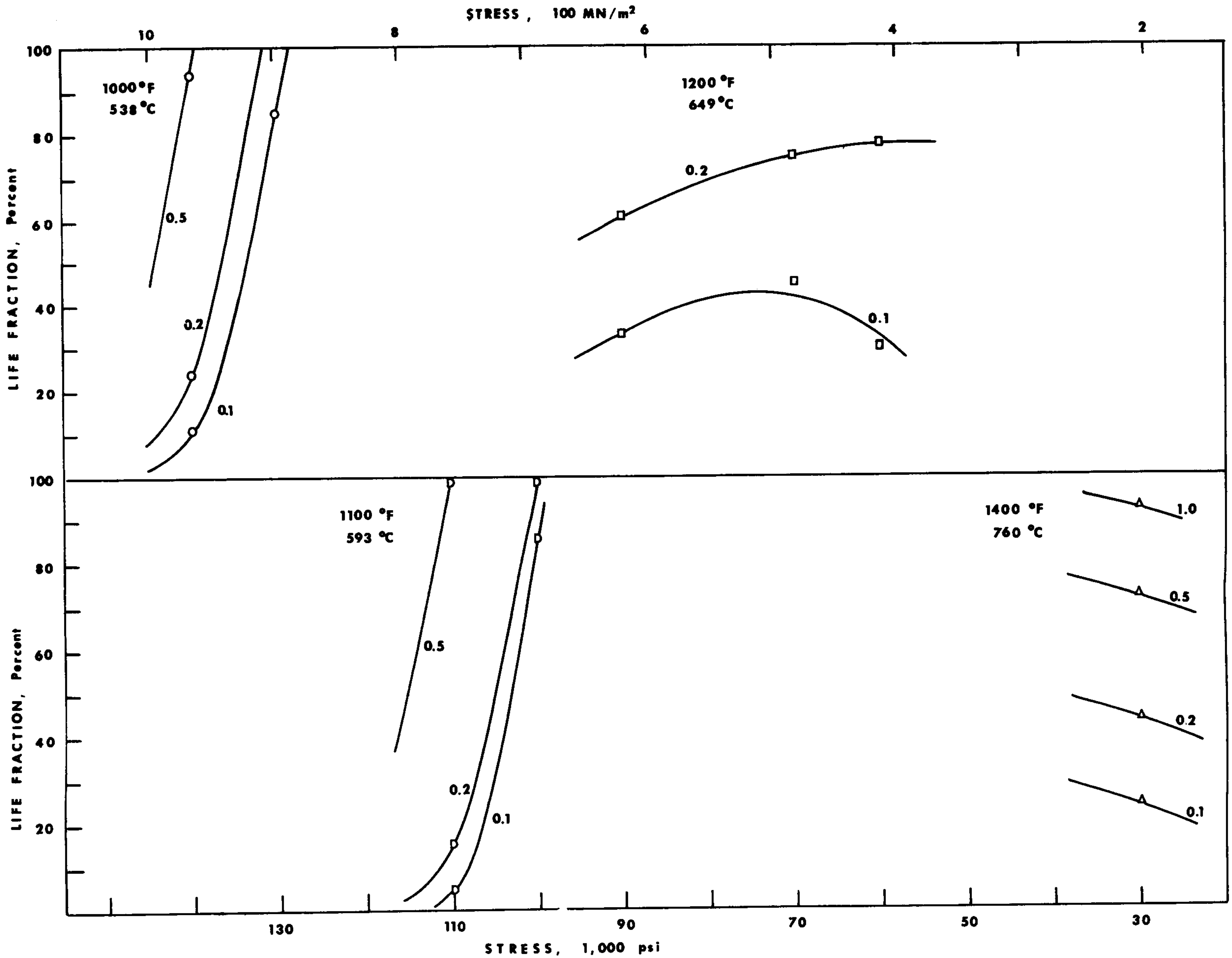


Figure 32. Iso-creep strain curves of life fraction versus stress at temperatures from 1000° to 1400°F (538 - 760°C) for Inconel 718 heat treated 1 hour at 1950°F plus 48 hours at 1350°F (732°C). Time-dependent notch sensitivity occurred under test conditions where large amounts of rupture life were utilized for small creep strains at test temperatures 1000°, 1100°, and 1200°F (538°, 593°, and 649°C).

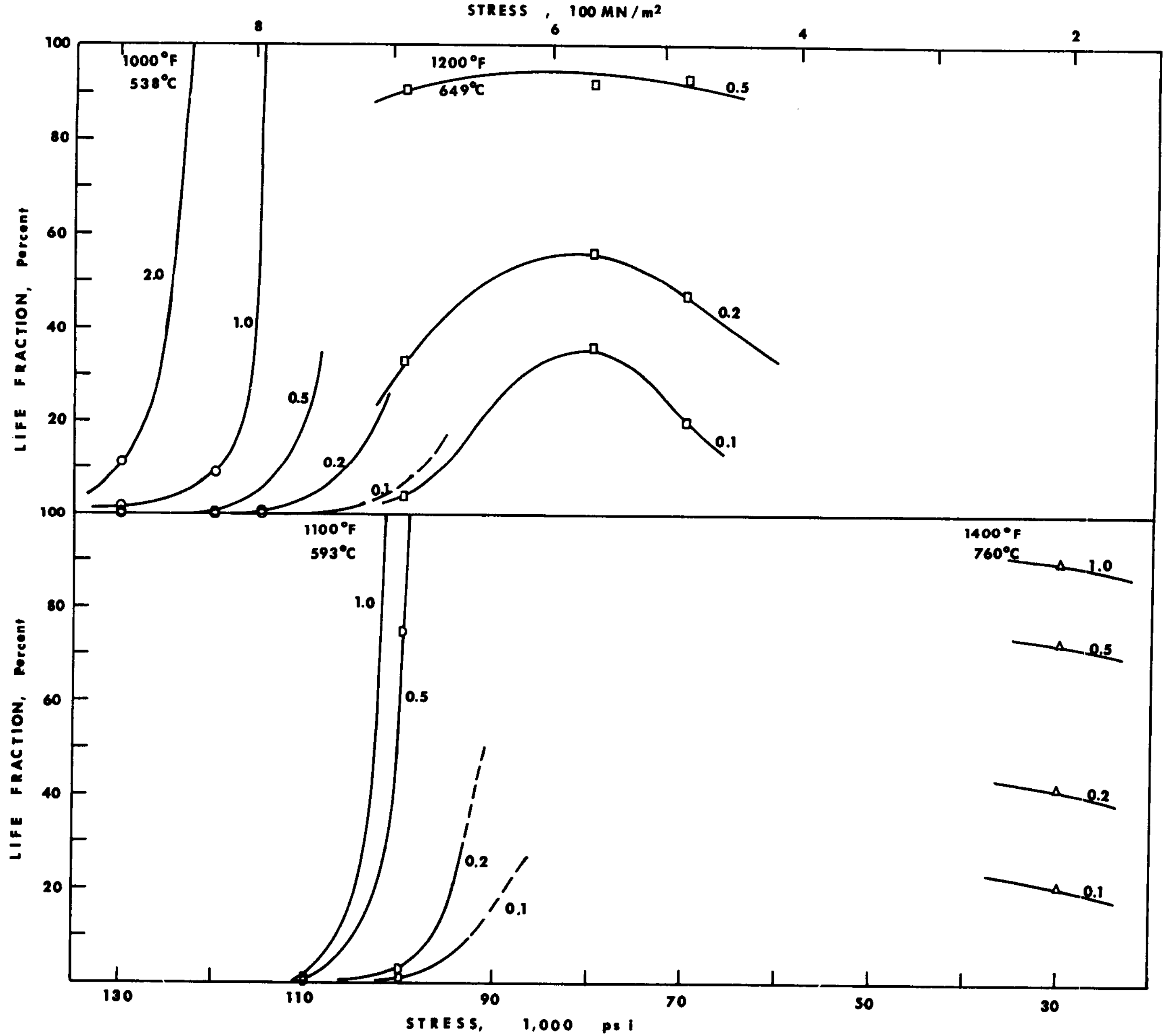


Figure 33. Iso-creep strain curves of life fraction versus stress at temperatures from 1000° to 1400°F (538 - 760°C) for Inconel 718 heat treated 1 hour at 1950°F (1066°C) plus 2 hours at 1550°F (843°C). Time-dependent notch sensitivity occurred under test conditions were large amounts of rupture life were utilized for small creep strains at test temperatures 1000° and 1100°F (538° and 593°C) (and to a lesser extent at 1200°F) (649°C).

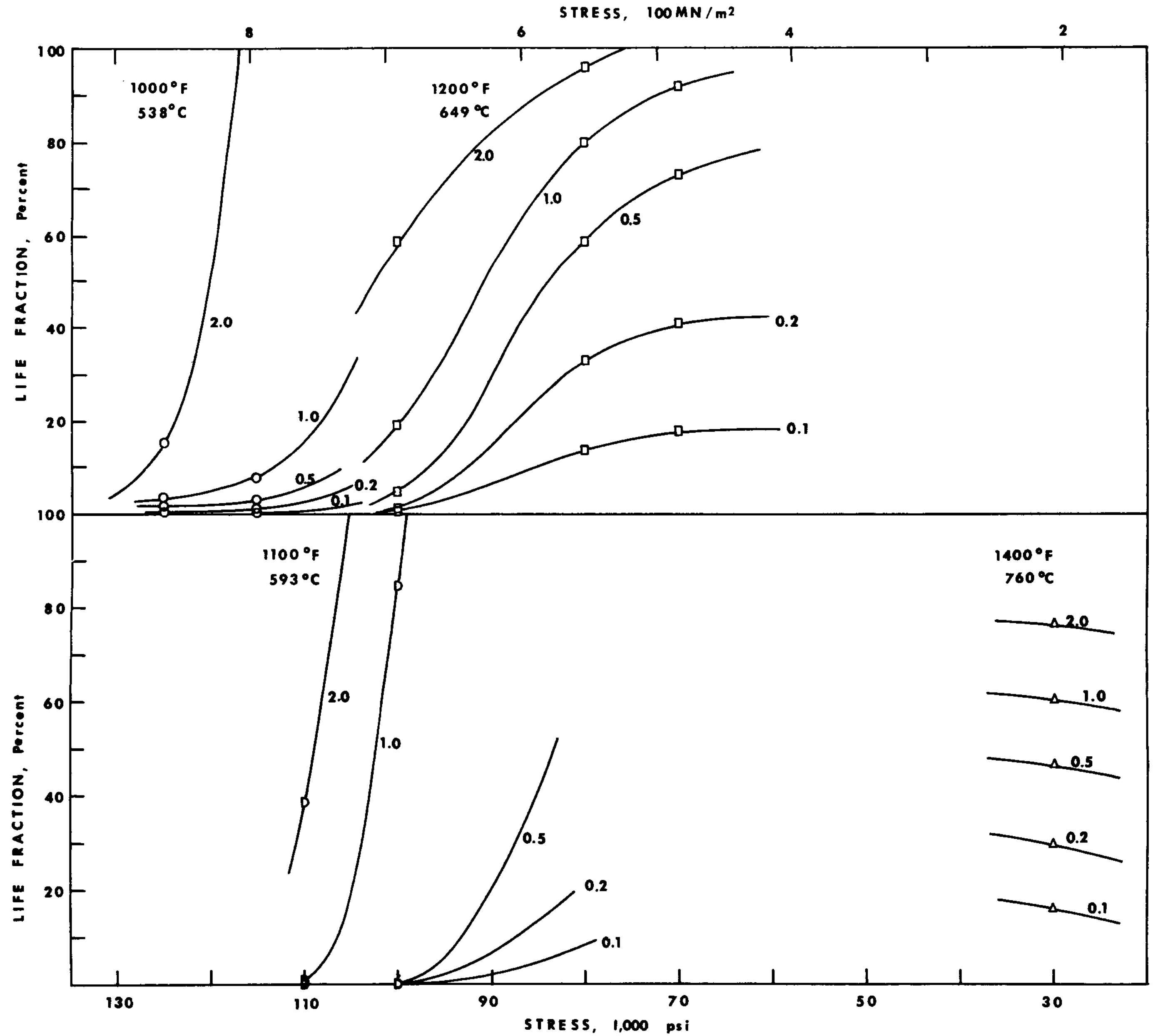


Figure 34. Iso-creep strain curves of life fraction versus stress at temperatures from 1000° to 1400° F (538 - 760°C) for Inconel 718 heat treated 1 hour at 1950°F (1066°C) plus 24 hours at 1550°F (843°C). For the test conditions evaluated (except possibly those at 1200°F (649°C)), the life fractures utilized for small amounts of creep were relatively low and no-time dependent notch sensitivity

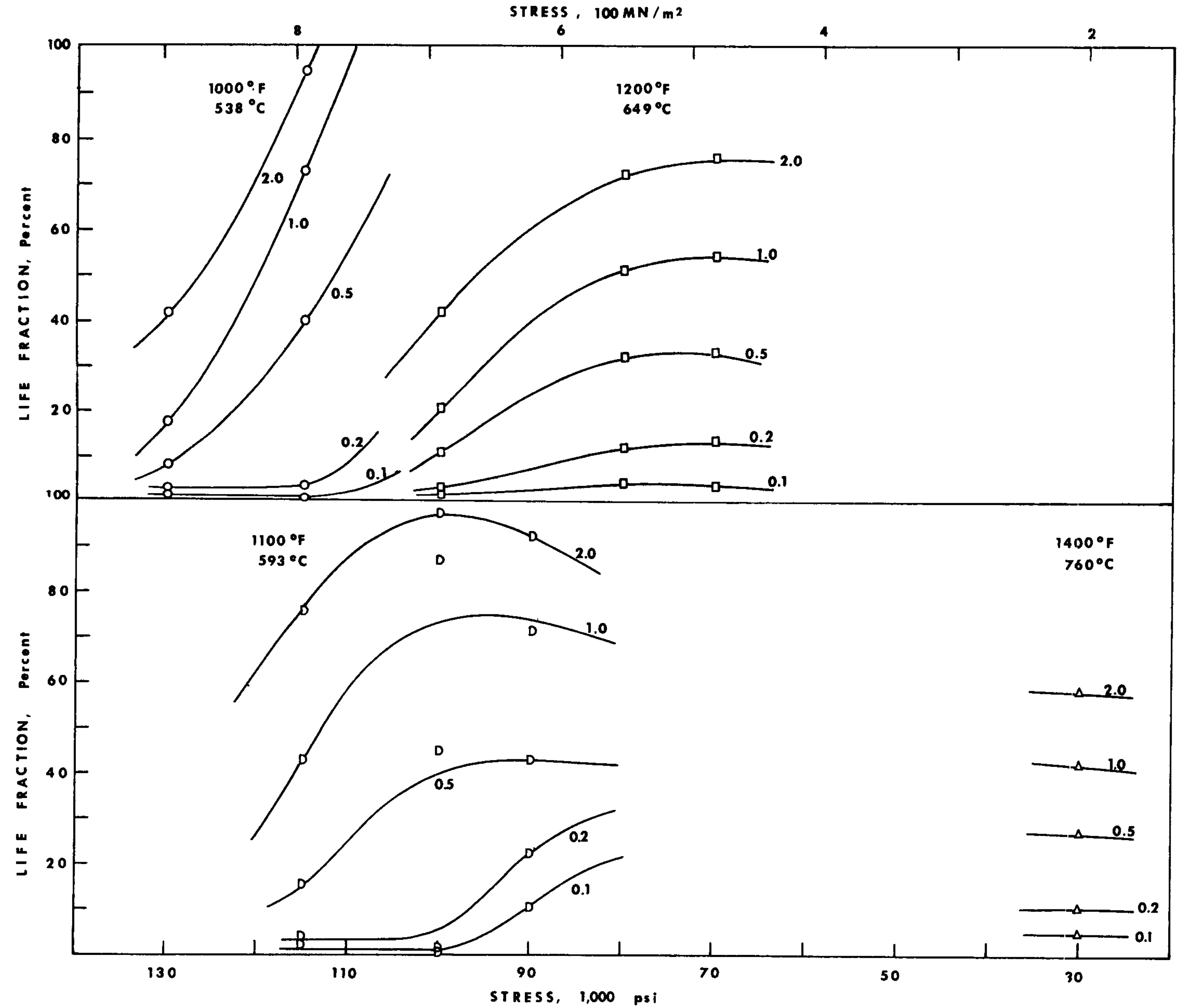


Figure 35. Iso-creep strain curves of life fraction versus stress at temperatures from 1000° to 1400°F (538 - 760°C) for Inconel 718 heat treated 10 hours at 1800°F (982°C) plus 48 hours at 1350°F (732°C). Time-dependent notch sensitivity occurred at 1000° and 1100°F (538 and 593°C) under test conditions where the life fractions utilized for small amounts of creep were high.

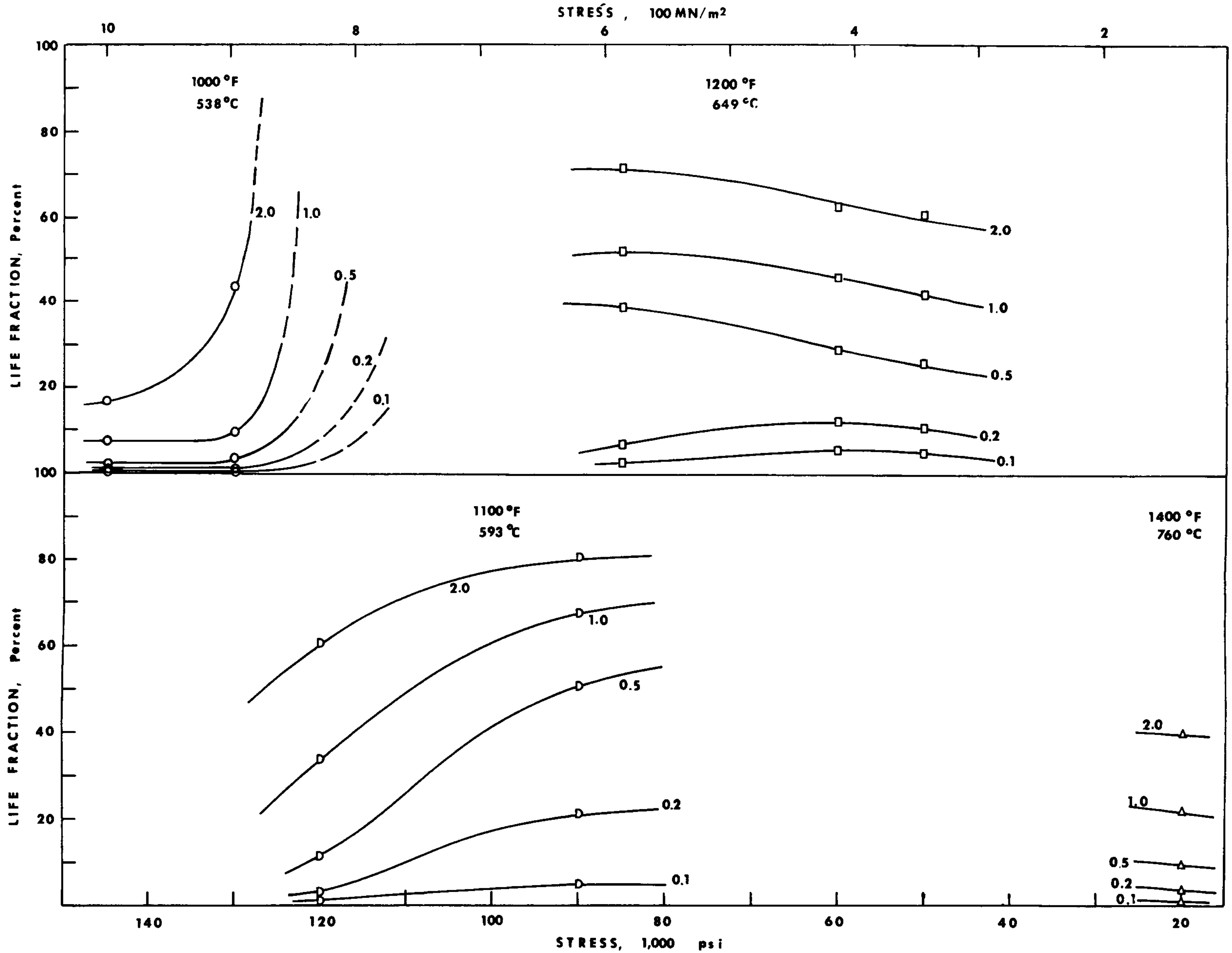


Figure 36. Iso-creep strain curves of life fraction versus stress at temperatures from 1000° to 1400°F (538 - 760°C) for Inconel 718 heat treated 10 hours at 1700°F (927°C) plus 3 hours at 1325°F (718°C). Time-dependent notch sensitivity occurred at 1000°F (538°C) for test conditions were large amounts of rupture life were utilized for small creep strains.

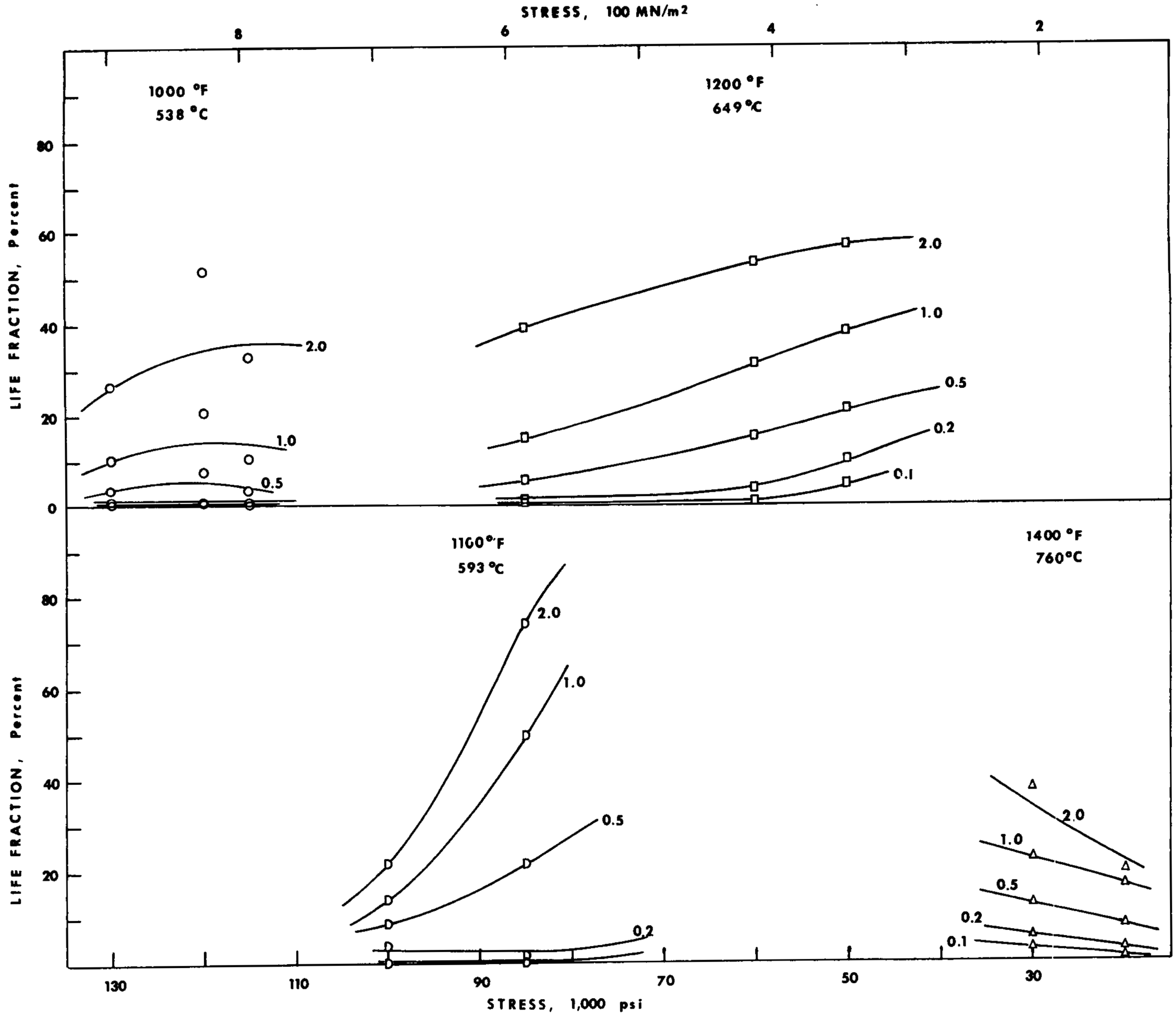


Figure 37. Iso-creep strain curves of life fraction versus stress at temperatures from 1000° to 1400°F (538 - 760°C) for Inconel 718 heat treated 10 hours at 1700°F (927°C) plus 48 hours at 1350°F (732°C). Little rupture life was utilized for small amounts of creep under all test conditions and no time-dependent notch sensitivity was observed.

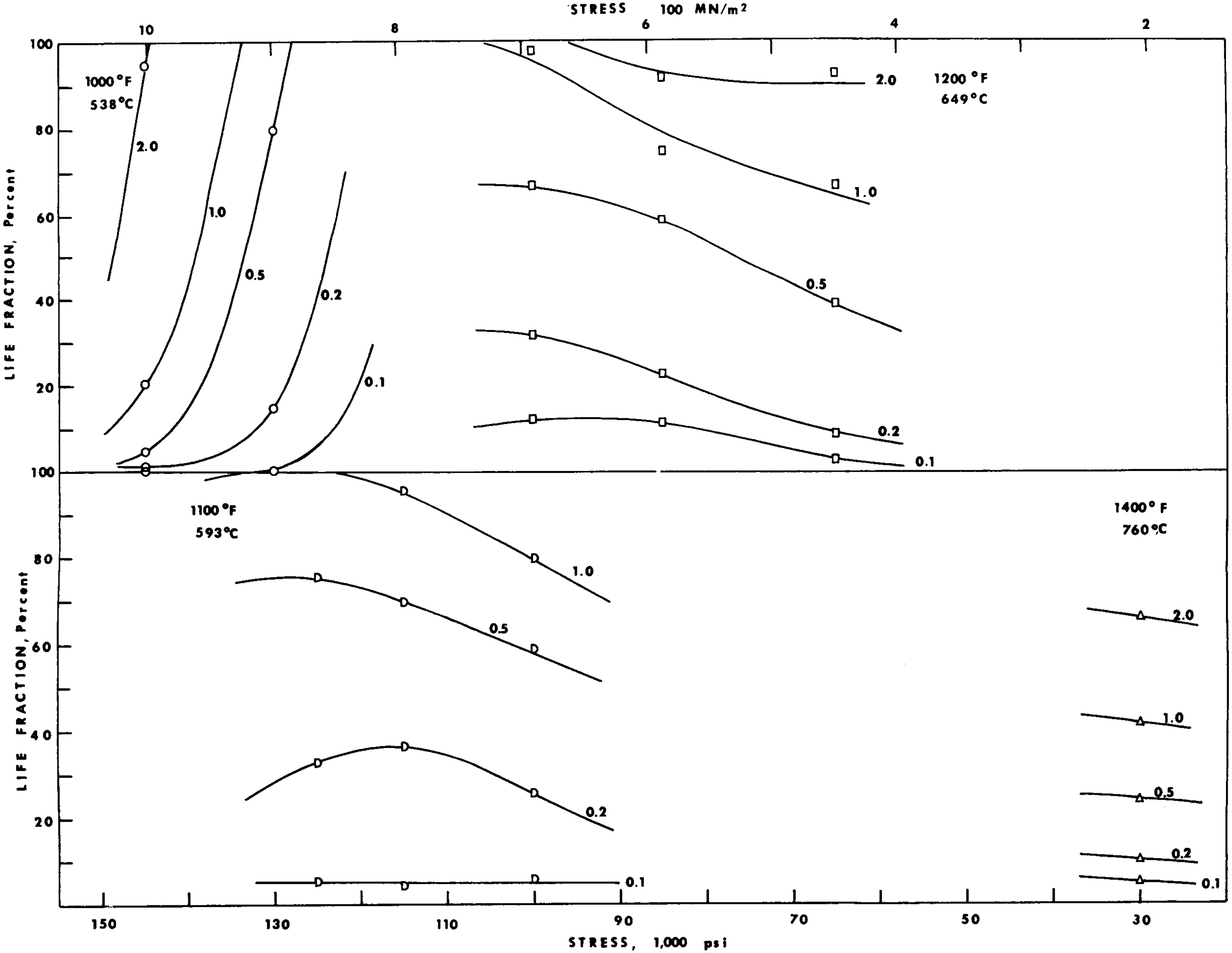


Figure 38. Iso-creep strain curves of life fraction versus stress at temperatures from 1000° to 1400°F (538 - 760°C) for Inconel 718 heat treated 1 hour at 1700°F (927°C) plus 3 hours at 1325°F (718°C). Time-dependent notch sensitivity was observed at 1000°F (538°C) for test conditions where the life fractions utilized for small amounts of creep were high.

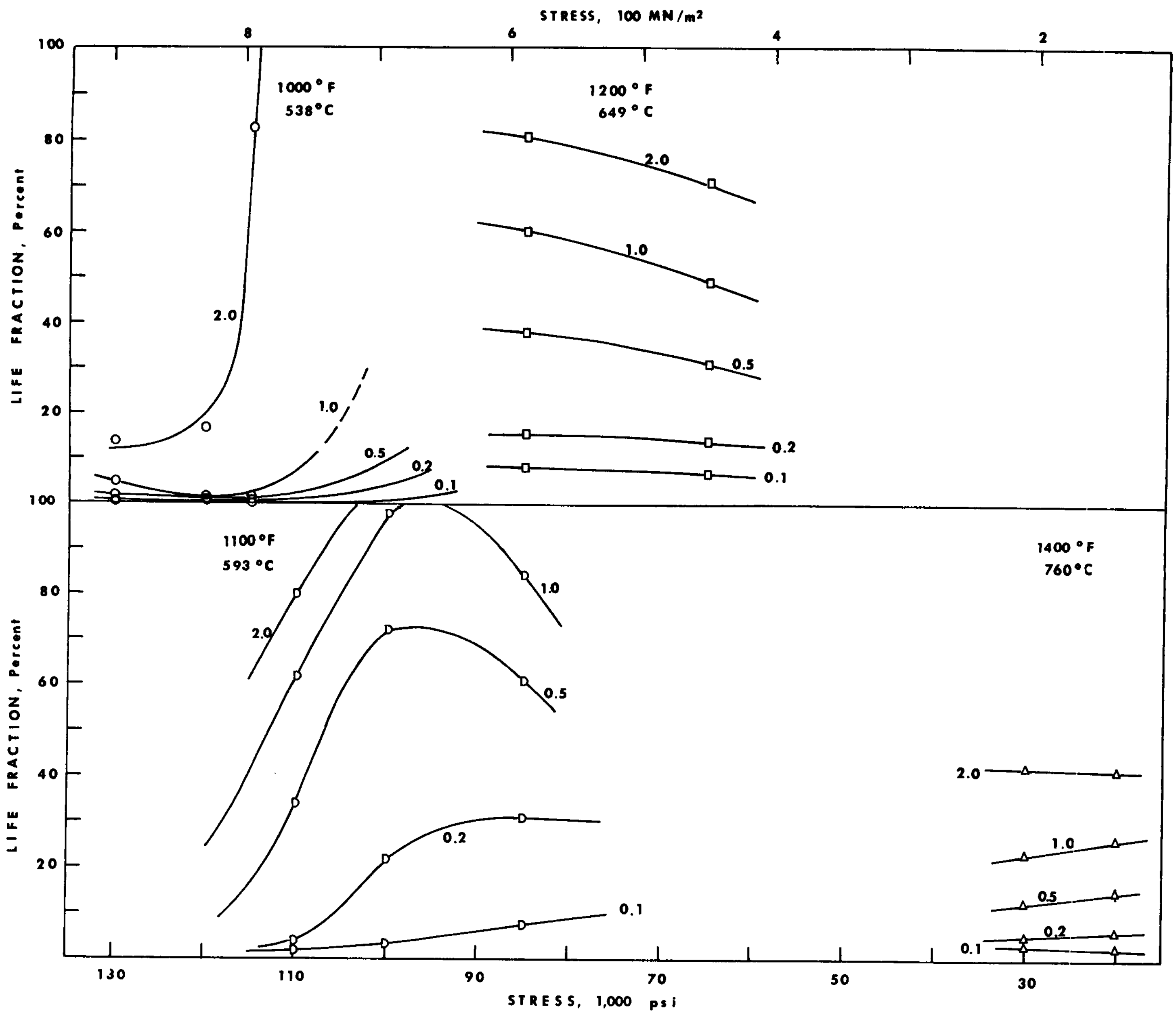
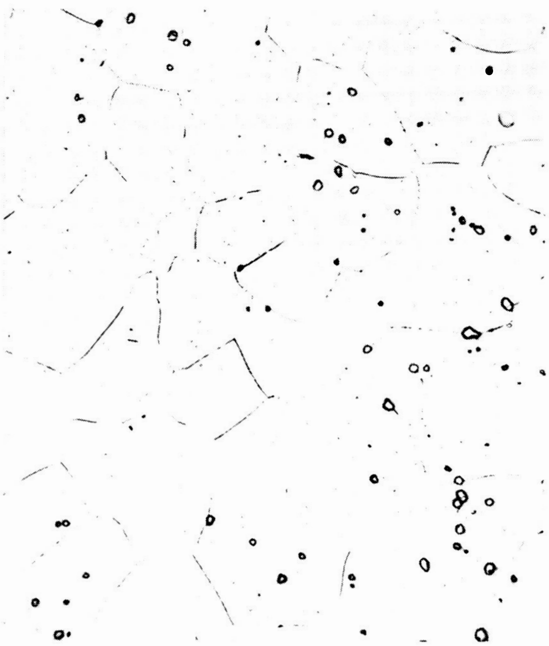
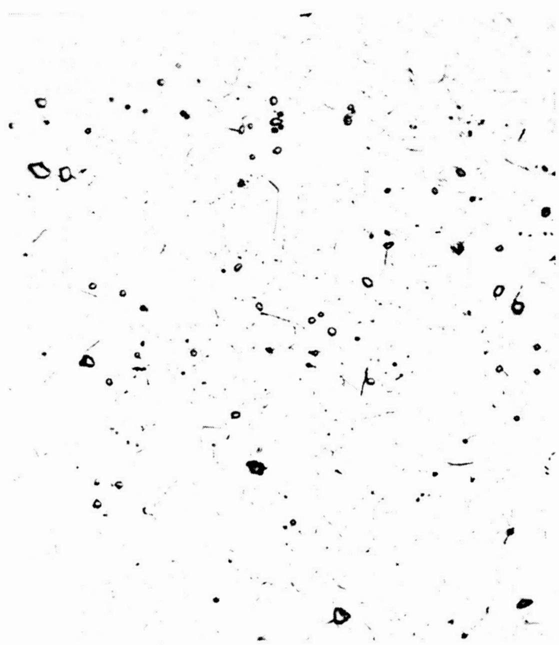


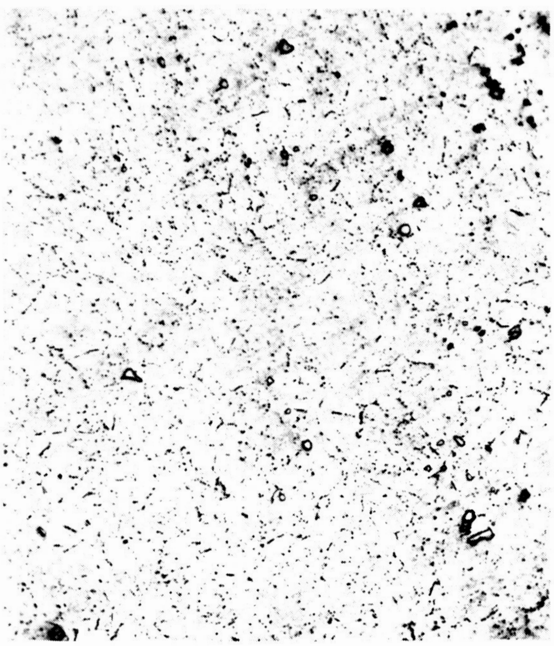
Figure 39. Iso-creep strain curves of life fraction versus stress at temperatures from 1000° to 1400°F (538 - 760°C) for Inconel 718 heat treated 1 hour at 1700°F (927°C) plus 2 hours at 1550°F (843°C). No time-dependent notch sensitivity was observed for all test conditions. The life fractions utilized for small amounts of creep were relatively low for all test conditions.



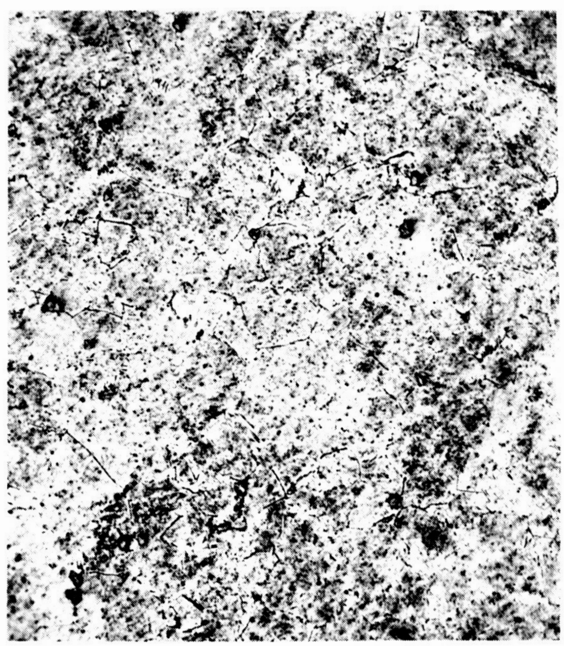
(a) 10 hrs. at 1950°F (1066°C) plus 48 hrs. at 1350°F(732°C)



(b) 1 hr. at 1950°F (1066°C) plus 48 hrs. at 1350°F (732°C)



(c) 1 hr. at 1950°F (1066°C) plus 2 hrs. at 1550°F (843°C)



(d) 1 hr. at 1950°F (1066°C) plus 24 hrs. at 1550°F(843°C)

Figure 40. Optical photomicrographs of Inconel 718 sheet showing differences in microstructure induced by variations in heat treatment. (250x)

(e) 10 hrs. at 1800°F (982°C)

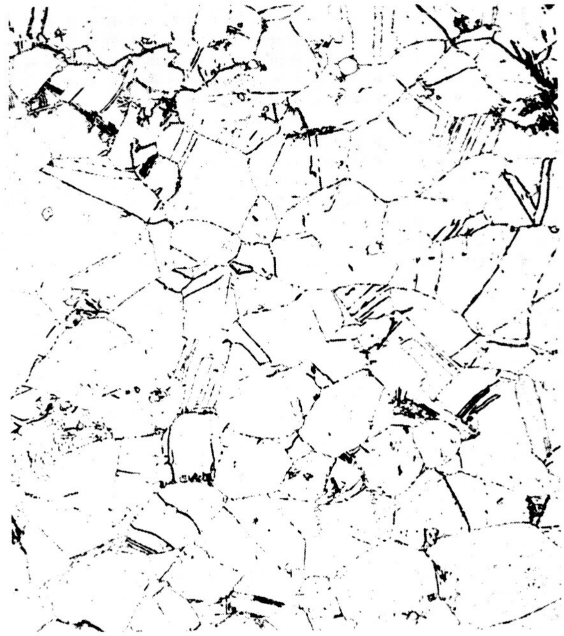
(f) 10 hrs. at 1800°F (982°C) plus 48 hrs. at 1350°F (732°C)

(g) 10 hrs.at 1700°F (927°C)

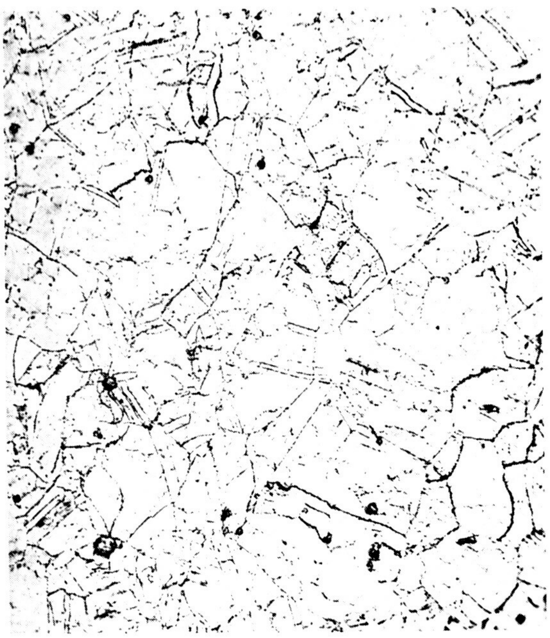
(h) 10 hrs. at 1700°F (927°C°) plus 3 hrs. at 1325°F (718°C)

Figure 40. (Continued) (250x)

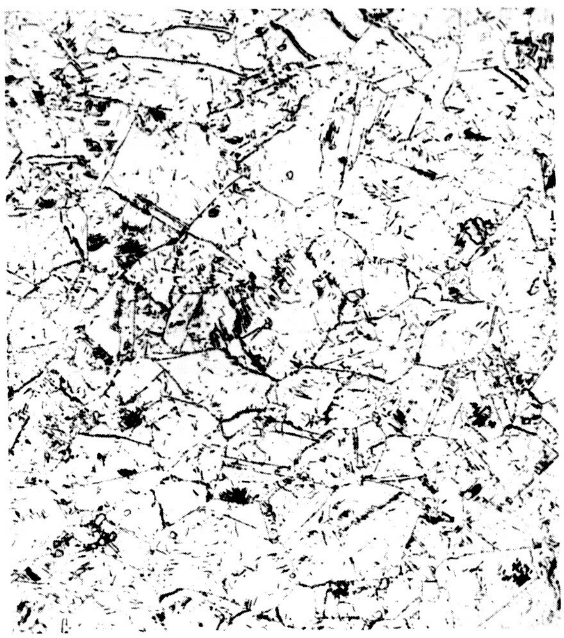
(i) 10 hrs. at 1700°F (927°C)
plus 48 hrs. at 1350°F (732°C)



(j) 1 hr. at 1700°F (927°C)



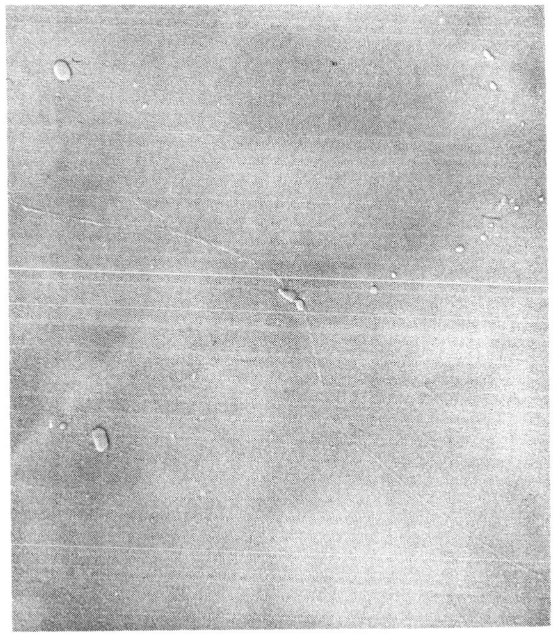
(k) 1 hr. at 1700°F (927°C)
plus 3 hrs. at 1325°F (718 °C)



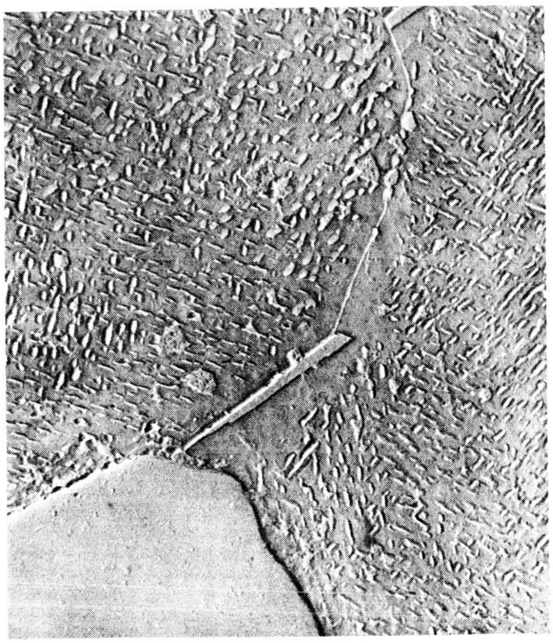
(l) 1 hr. at 1700°F (927°C)
plus 2 hrs. at 1550°F (843°C)

Figure 40. (Continued) (250x)

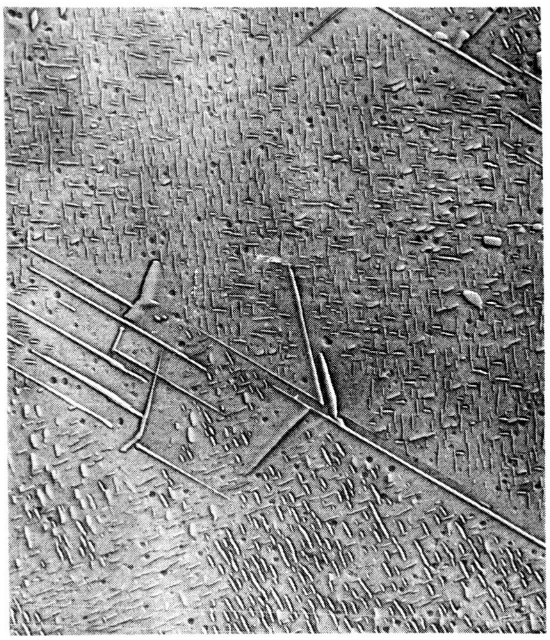
(a) 10 hrs. at 1950°F (1066°C) plus 48 hrs. at 1350°F (732°C)



(b) 1 hr. at 1950°F (1066°C) plus 48 hrs. at 1350°F (732°C)

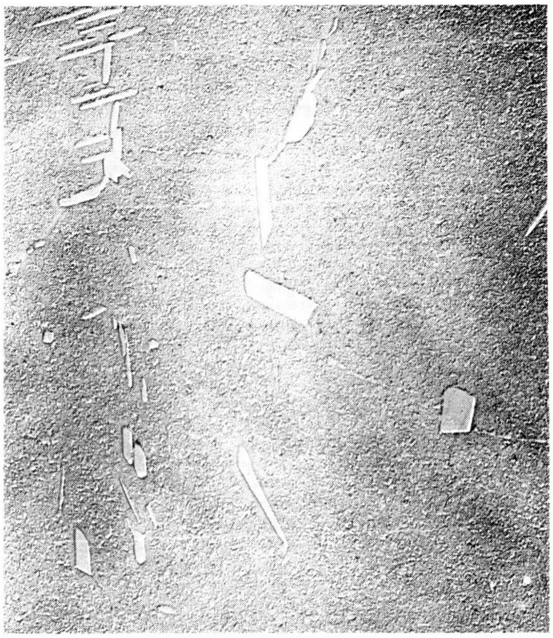


(c) 1 hr. at 1950°F (1066°C) plus 2 hrs. at 1550°F (843°C)

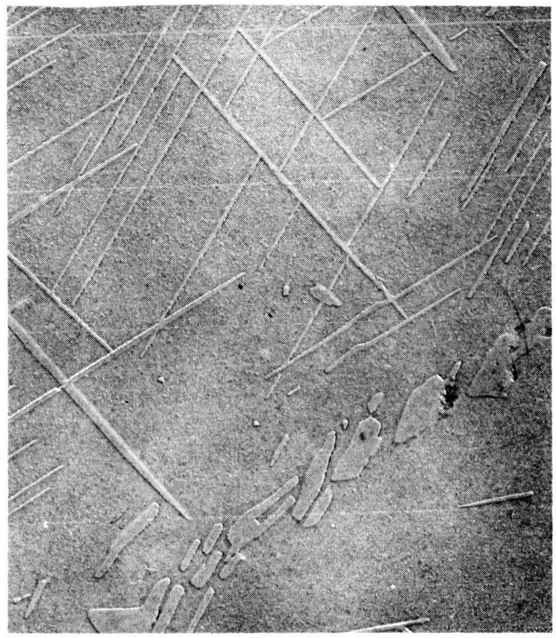


(d) 1 hr. at 1950°F (1066°C) plus 24 hrs. at 1550°F (843°C)

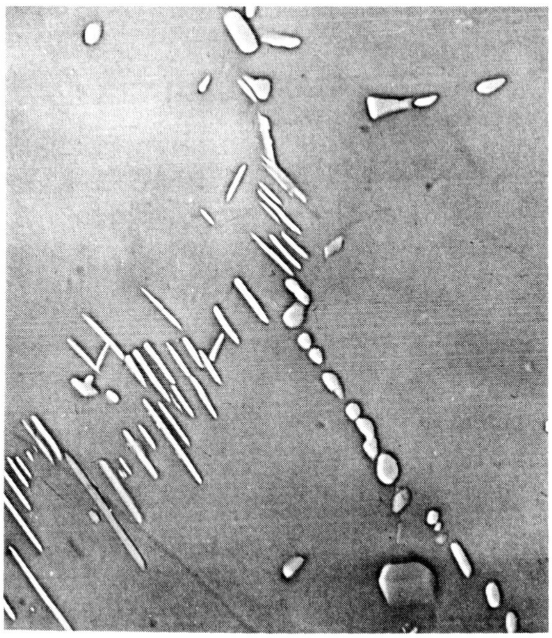
Figure 41. Replica electron micrographs of Inconel 718 sheet in the as-heat treated conditions showing variations in microstructural features. (6000X)



(e) 10 hrs. at 1800°F (982°C) plus 48 hrs. at 1350°F (732°C)



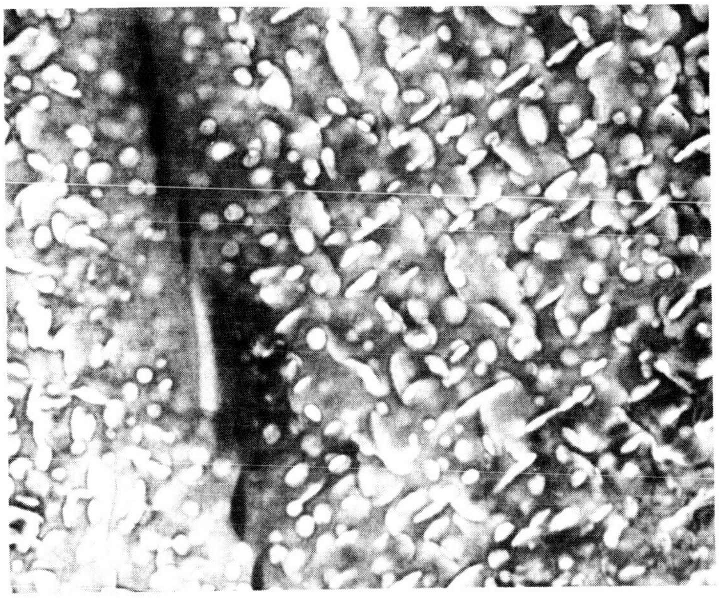
(f) 10 hrs. at 1700°F (927°C) plus 48 hrs. at 1350°F (732°C)



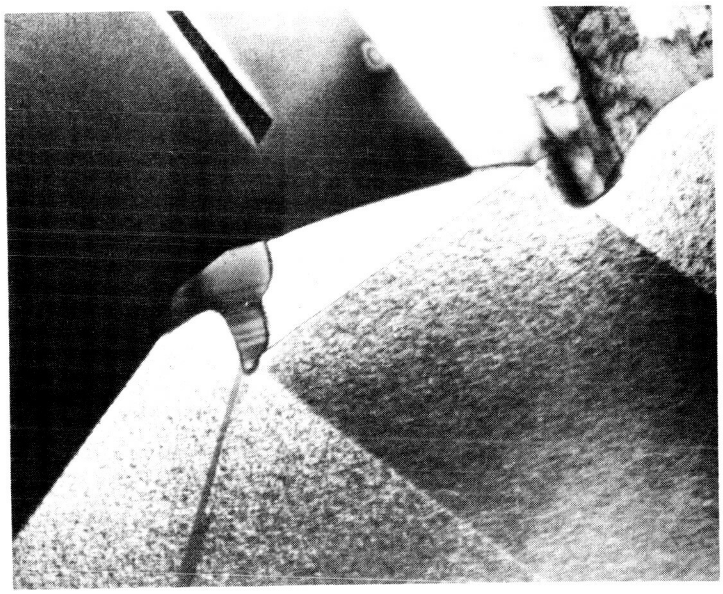
(g) 1 hr. at 1700°F (927°C) plus 3 hrs. at 1325°F (718°C)

(h) 1 hr. at 1325°F (718°C) plus 2 hrs. at 1550°F (843°C)

Figure 41. (Continued) (6000X)



(a) 1 hr. at 1950°F(1066°C) plus 48 hrs. at 1350°F (732°C) 85,000x



(b) 1 hr. at 1700°F (927°C) plus 3 hrs. at 1325°F (718°C) 30,000x

Figure 42. Transmission electron micrographs of Inconel 718 in two heat treated conditions showing γ'/γ'' size distributions.

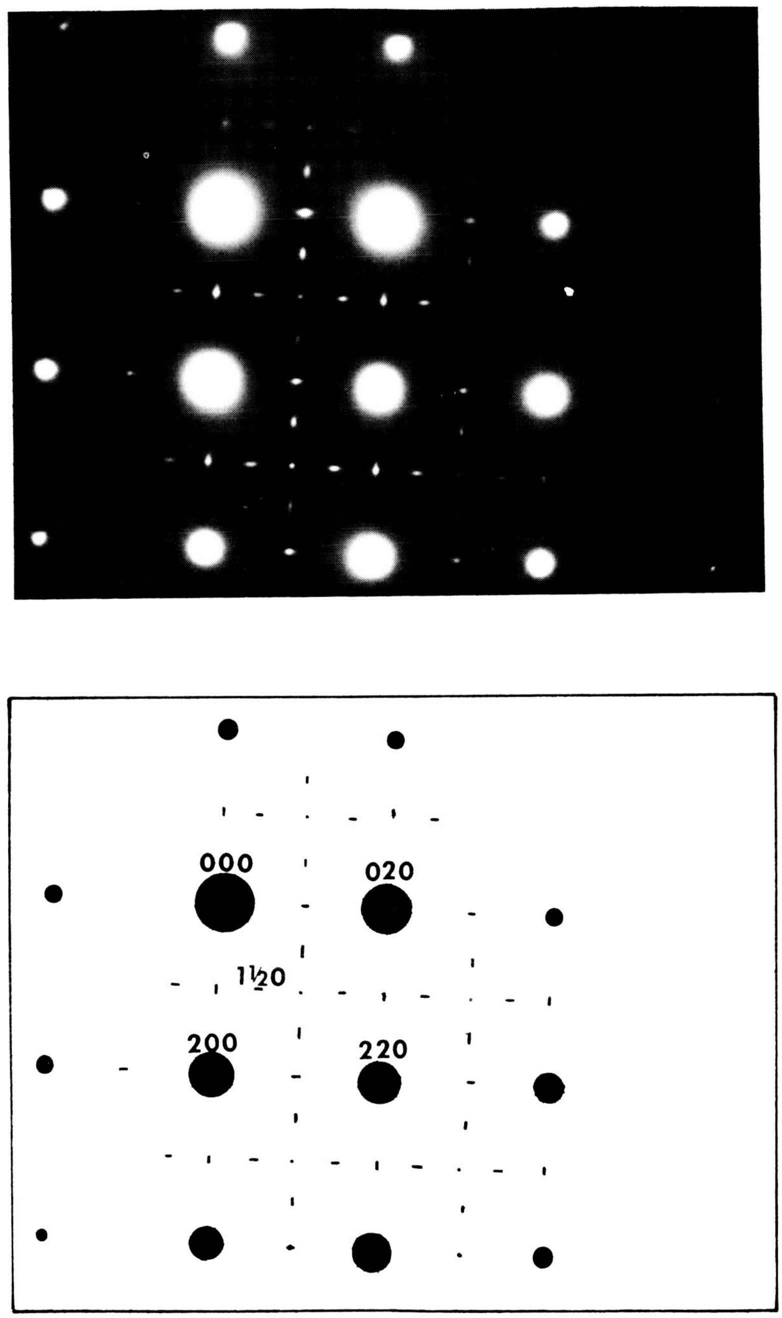
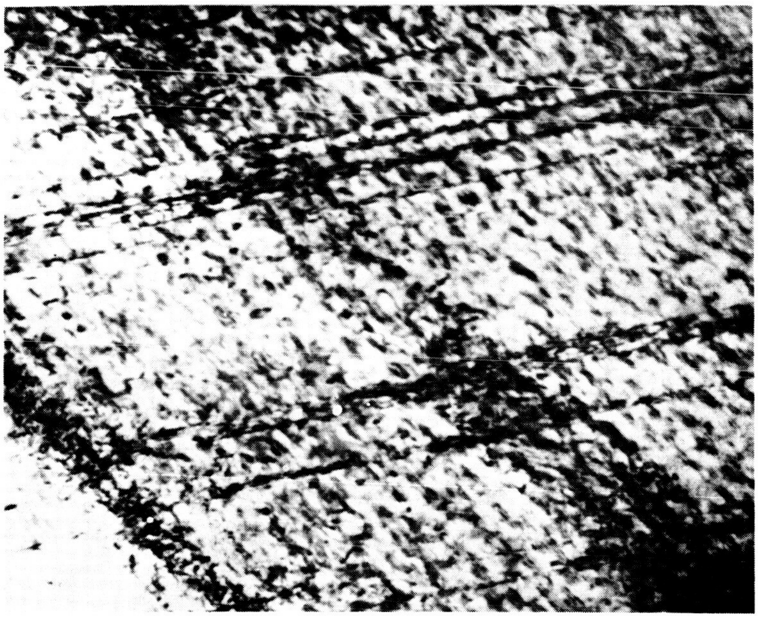
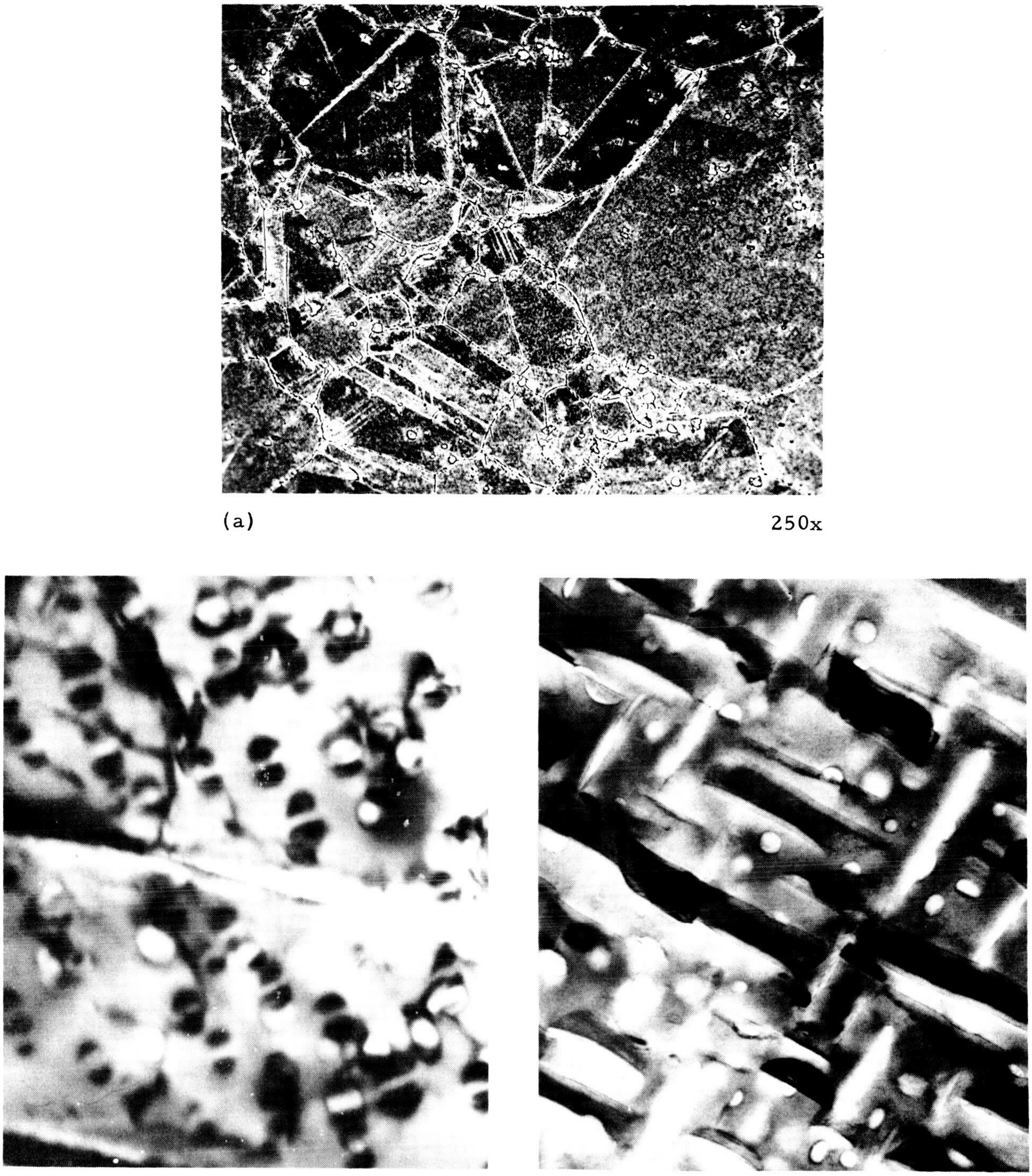


Figure 43. Thin foil diffraction pattern. [100] matrix zone. The superlattice spots arise from γ'' with three orientations.



50,000x

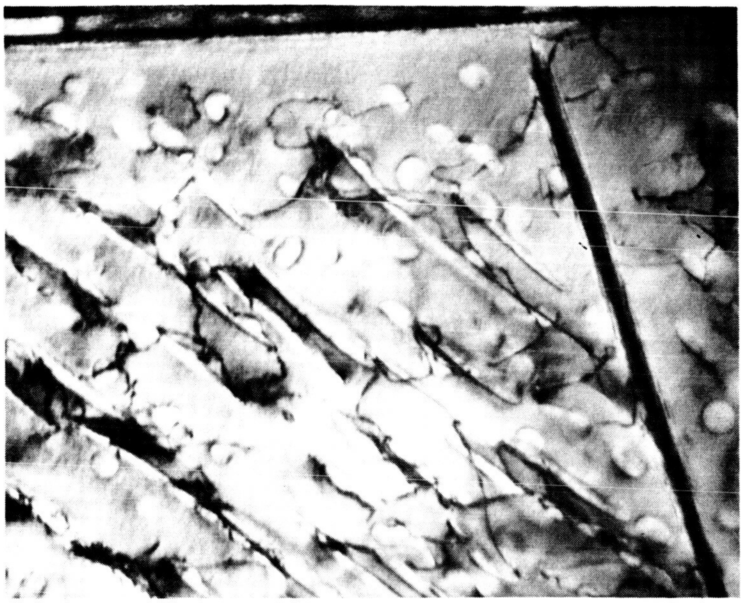
Figure 44. Thin-foil electron micrograph of Inconel 718, heat treated 1 hour at 1950°F(1066°C) plus 48 hours at 1350°F (732°C) and creep-rupture tested at 120ksi (827MN/m^2) at 1100°F (593°C) (ruptured in 1.4 hours at 4.2% elongation). Bands are evident that resulted from localized deformation.



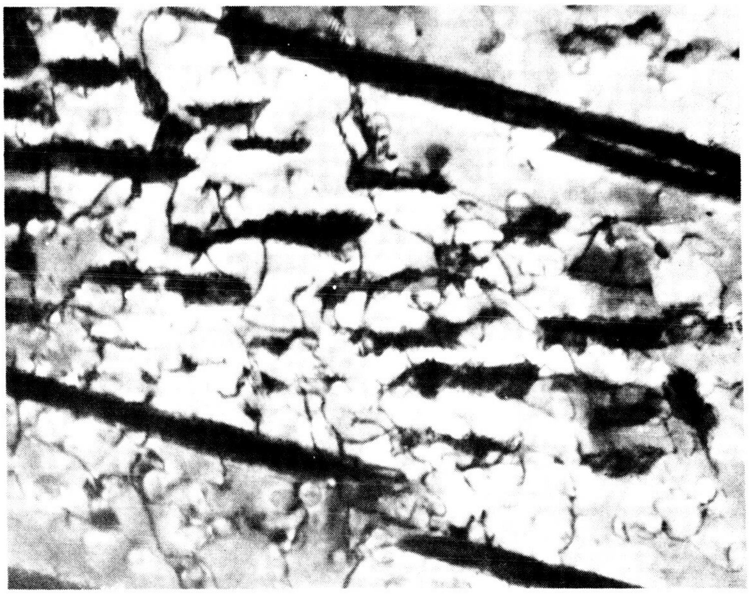
(a) 250x

(b) 65,000x (c) 40,000x

Figure 45. Optical and transmission electron micrographs of Inconel 718 heat treated 1 hour at 1950°F (1066°C) plus 48 hours at 1350°F (732°C) and creep rupture tested at 30ksi(207MN/m^2) at 1400°F(760°C) (ruptured in 384 hours at 2.1% elongation). During the test exposure Ni_3Cb precipitated and the γ' and γ'' increased in size considerably. Contrast effects associated with coherent γ' and γ'' are evident in (b) and (c).



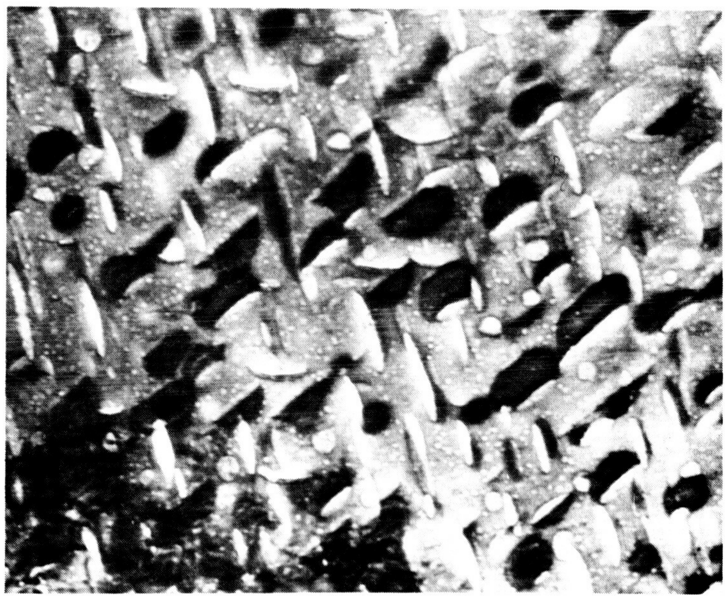
(a) 45,000x



(b) 40,000x

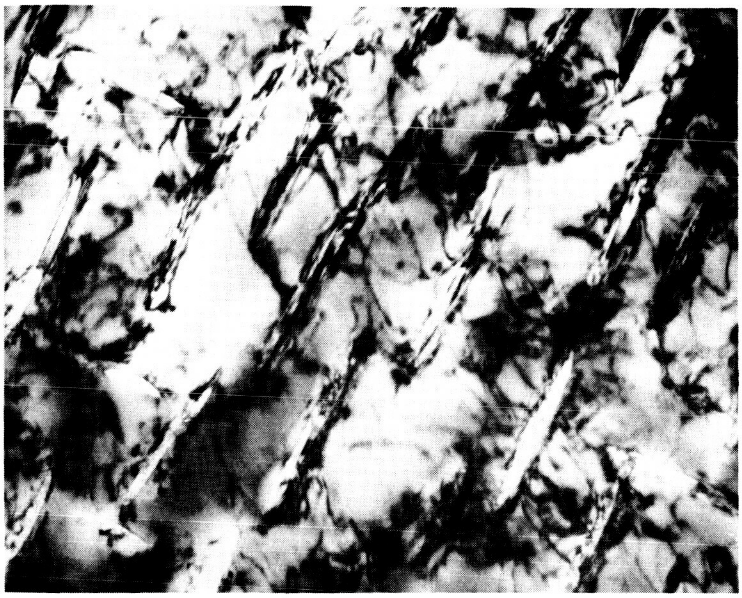
Figure 46. Thin-foil electron micrographs of Inconel 718 heat treated 1 hour at 1950°F(1066°C) plus 48 hours at 1350°F(732°C) and creep-rupture tested at 30ksi(207MN/m^2) at 1400°F(760°C). The deformation is homogeneous. Dislocation can be observed entangled with the γ'' precipitate and bowing between γ' particles leaving "pinched off" dislocation loops.

(a) 60,000x

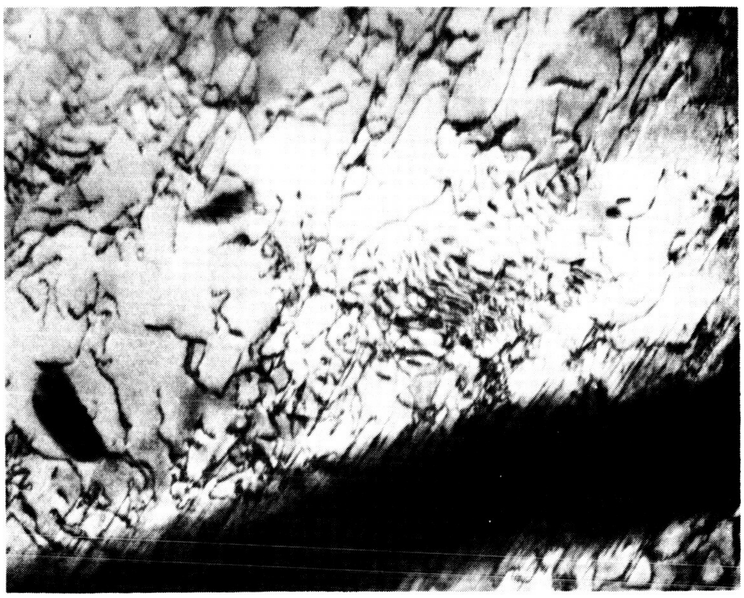


(b) 75,000x

Figure 47. Transmission electron micrographs of Inconel 718 heat treated 1 hour at 1950°F (1066°C) plus 2 hours at 1550°F (843°C) and tested at 100ksi($690 MN/m^2$) at 1100°F(593°C) (ruptured in 385 hours). The deformation in (a) is localized. In (b) a fine dispersion of γ'/γ'' is evident which presumably developed during the test exposure.



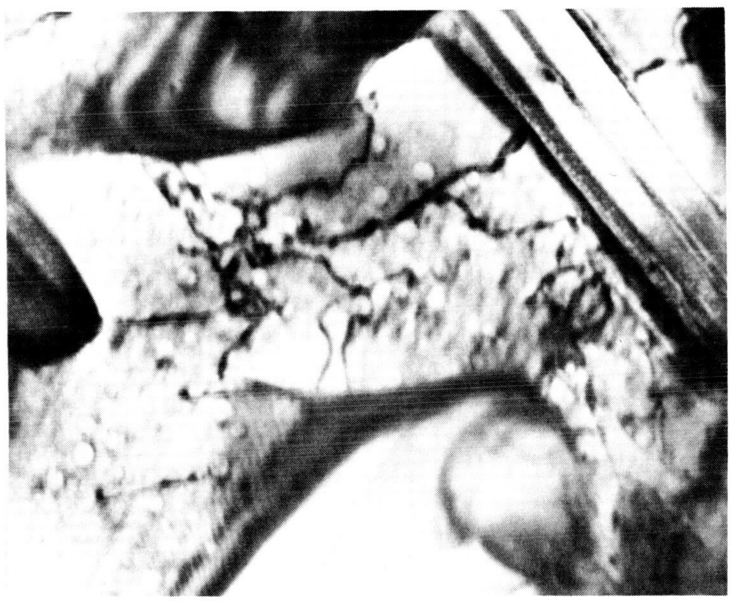
(a) 60,000x



(b) 30,000x

Figure 48. Thin-foil electron micrographs of Inconel 718 heat treated 1 hour at 1950°F (1066°C) plus 24 hours at 1550°F (843°C) and tested at 115ksi(793MN/m^2) at 1000°F (538°C) (rupture in 1857 hours at 10.2% elongation). The deformation is homogeneous.

(a) 20,000x



(b) 100,000x

Figure 49. Transmission electron micrograph of Inconel 718 heat treated 1 hour at 1700°F(927°C) plus 3 hours at 1325°F(718°C) and creep rupture tested-(a) at 1000°F(538°C) at 130ksi(896MN/m^2) (ruptured in 5613 hours at 3.5% elongation), (b) at 1200°F (649°C) at 65ksi(448MN/m^2) (ruptured in 937 hours at 3.7% elongation). In the lower temperature test dislocations, extended to form stacking fault ribbons, sheared the γ'/γ'' precipitates. During the higher temperature test exposure γ'/γ'' growth occurred. In consequence, the deformation was homogeneous and the dislocations by-passed the precipitate particles.

70,000x

Figure 50. Thin foil electron micrograph of Inconel 718 heat treated 10 hours at 1700°F (927°C) plus 48 hours at 1350°F (732°C) and creep-rupture tested at 1000°F (538°C) at 120ksi ($827 MN/m^2$) (ruptured in 1382 hours at 5.6% elongation). The deformation is homogeneous The dislocations can be observed bowing between precipitate particles leaving "pinched off" dislocation loops.

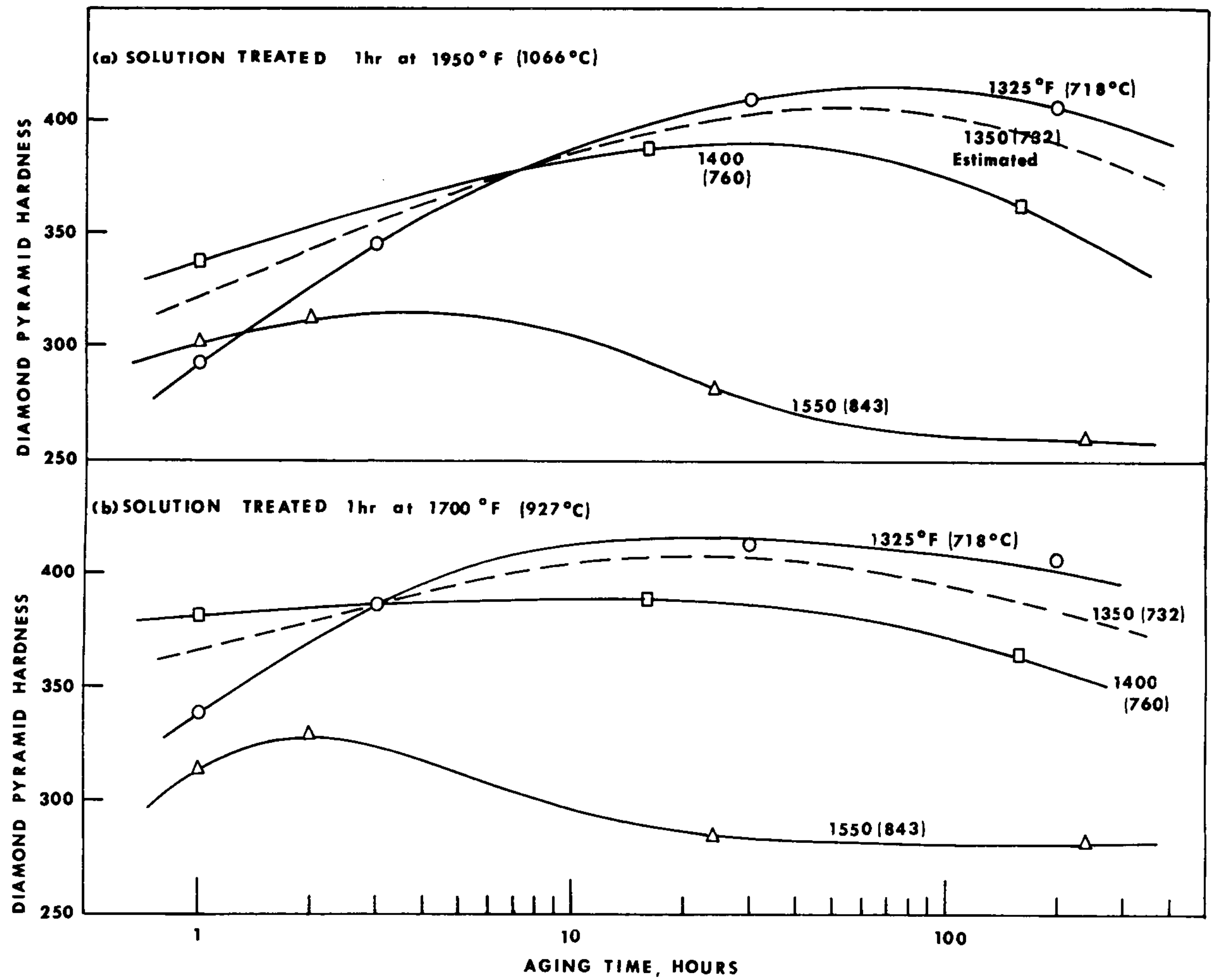


Figure 51. Effect of aging exposures at 1325°, 1400° and 1550°F (718, 760 and 843°C) on the Diamond Pyramid Hardness of several heat treatments of 0.030-inch (.75mm) thick Inconel 718 sheet.

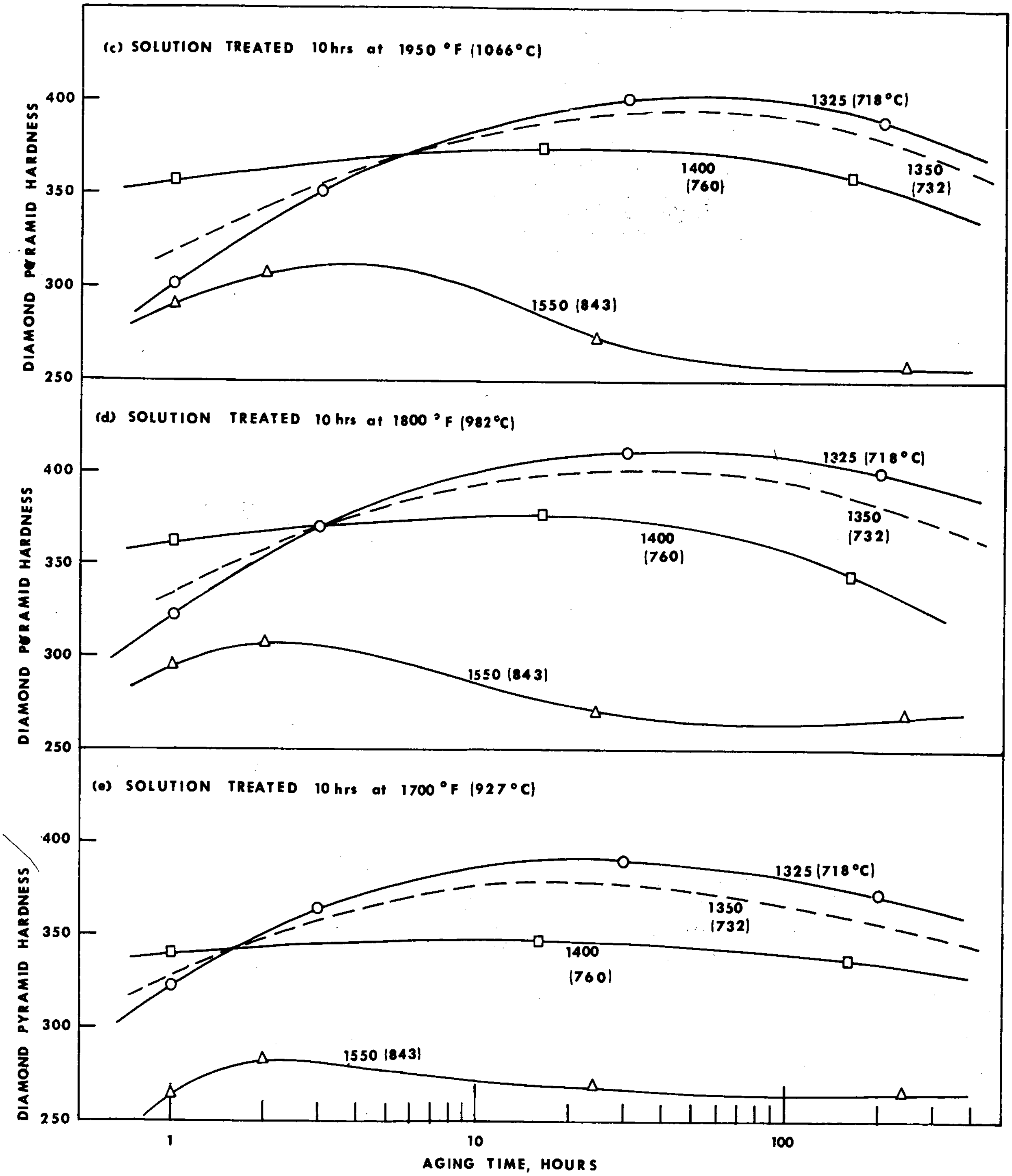


Figure 51. (continued)